STAR (1+26)

NASA Technical Memorandum 83006

COSAM Program Overview

Conservation Of Strategic Aerospace Materials

Material presented at a government-industry-university information exchange workshop held at NASA Lewis Research Center, Cleveland, Ohio, October 14-15, 1982

October 1982

NASA

PREFACE

The United States is highly dependent on foreign sources for many materials required for its economic health. In the aerospace industry the four metals chromium, cobalt, columbium, and tantalum have been identified as strategic materials. The United States imports in excess of 90 percent of each of these metals, and one foreign country currently controls a major portion of the U.S. supply. The National Materials and Minerals Policy, Research, and Development Act of 1980 has helped to focus attention on this critical problem that faces not only the aerospace industry, but most other industries as well. Government agencies are responding to this Act by conducting research, holding public workshops and conferences, and coordinating efforts through committees such as COMAT.

The COSAM Program was initiated in 1980 with its formulation being carried out in cooperation with the aerospace community, in particular the aircraft engine industry. The program emphasizes the cooperative efforts at NASA Lewis Research Center, universities, and industry, and is focused on three aspects of the strategic materials problem:

- Creation of the needed understanding of the roles of strategic elements in nickel-base superalloys so as to allow their reduction through substitution by less strategic elements.
- Identification of ways to exploit advanced materials processing concepts for a similar goal.
- On a higher-risk, longer-term basis, identification of alternate materials with no strategic element content.

To provide representatives from government, industry, and universities the latest findings in the COSAM Program, a 2-day review was held in October 1982. This publication contains abstracts and figures of the presentations at that review.

Joseph R. Stephens
Manager, COSAM

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COSAM PROGRAM OVERVIEW

Joseph R. Stephens
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

NASA Lewis Research Center has undertaken a long-range program in support of the aerospace industry aimed at reducing the need for strategic materials used in gas turbine engines. The program is called "COSAM - Conservation Of Strategic Aerospace Materials." This program has three general objectives; they are

(1) To contribute basic scientific understanding to the turbine engine "technology bank" so as to maintain our national security in possible times of constriction or interruption of our strategic material supply lines.

(2) To help reduce the dependence of United States military and civilian gas turbine engines on disruptive worldwide supply/price fluctuations in regard to strategic materials.

(3) To help minimize the acquisition costs as well as optimize performance of such engines so as to contribute to the United States position of preeminence in world gas turbine engine markets.

To achieve these objectives, the COSAM program is developing the basic understanding of the roles of strategic elements in today's nickel-base superalloys and will provide the technology base upon which their use in future aircraft engine alloys/components can be decreased. Technological thrusts in three major areas are underway to meet these objectives. These thrusts consist of strategic element substitution, advanced processing concepts, and alternate material identification. Based on criticality of need, initial efforts are concentrated on the strategic elements of cobalt (97 percent imported), tantalum (91 percent imported), columbium (100 percent imported), and chromium (91 percent imported). The following is an overview of the COSAM Program and of the research projects that have been initiated to date.

ORIGINAL FAGE IS
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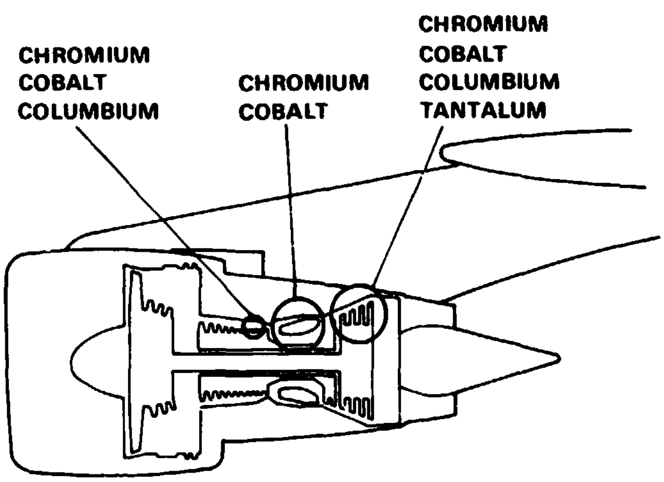
CONSERVATION OF STRATEGIC AEROSPACE MATERIALS

STRATEGIC MATERIALS

- DEFINED: THOSE PREDOMINANTLY OR WHOLLY IMPORTED ELEMENTS CONTAINED IN THE METALLIC ALLOYS USED IN AEROSPACE COMPONENTS WHICH ARE ESSENTIAL TO THE STRATEGIC ECONOMIC HEALTH OF THE U. S. AEROSPACE INDUSTRY
- IDENTIFIED: CHROMIUM
  COBALT
  COLUMBIUM
  TANTALUM

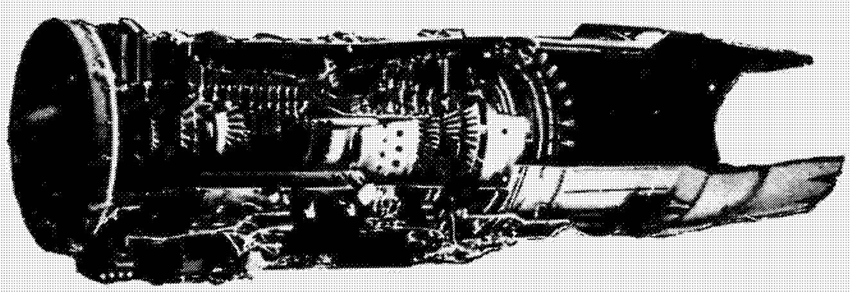
STRATEGIC METALS ARE CRITICAL TO TURBINE ENGINES



NEEDED FOR PERFORMANCE AND LONG LIFE

COBALT – HIGH TEMPERATURE STRENGTHENER
COLUMBIUM – INTERMEDIATE TEMPERATURE STRENGTHENER
TANTALUM – OXIDATION RESISTANCE
CHROMIUM – CORROSION RESISTANCE

F100 ENGINE INPUT REQUIREMENTS IN POUNDS



CHROMIUM	1485	COLUMBIUM	145
COBALT	885	TANTALUM	3

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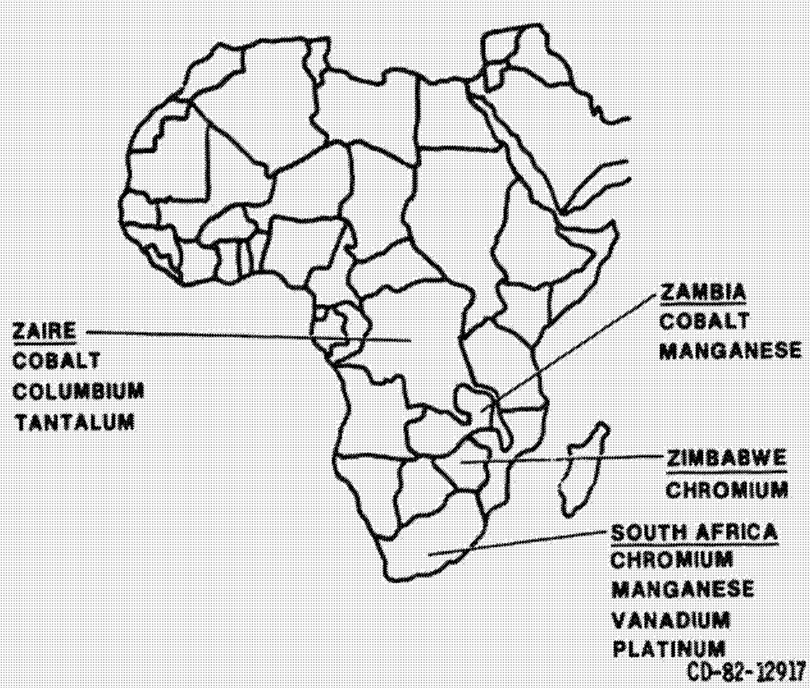
STRATEGIC MATERIAL RESOURCES IN AFRICA



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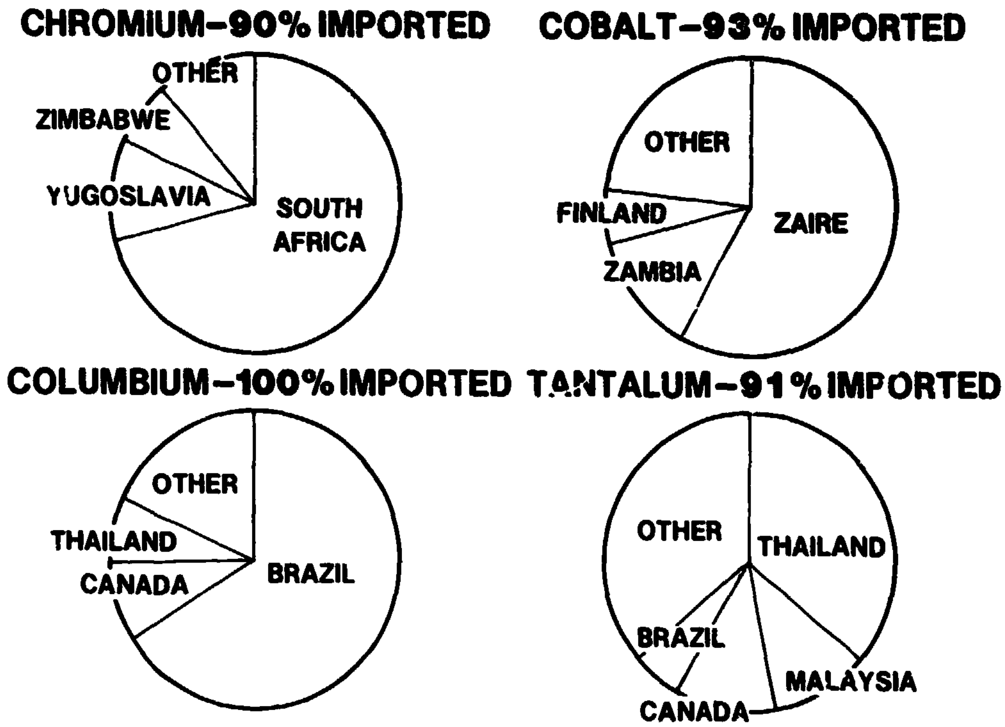
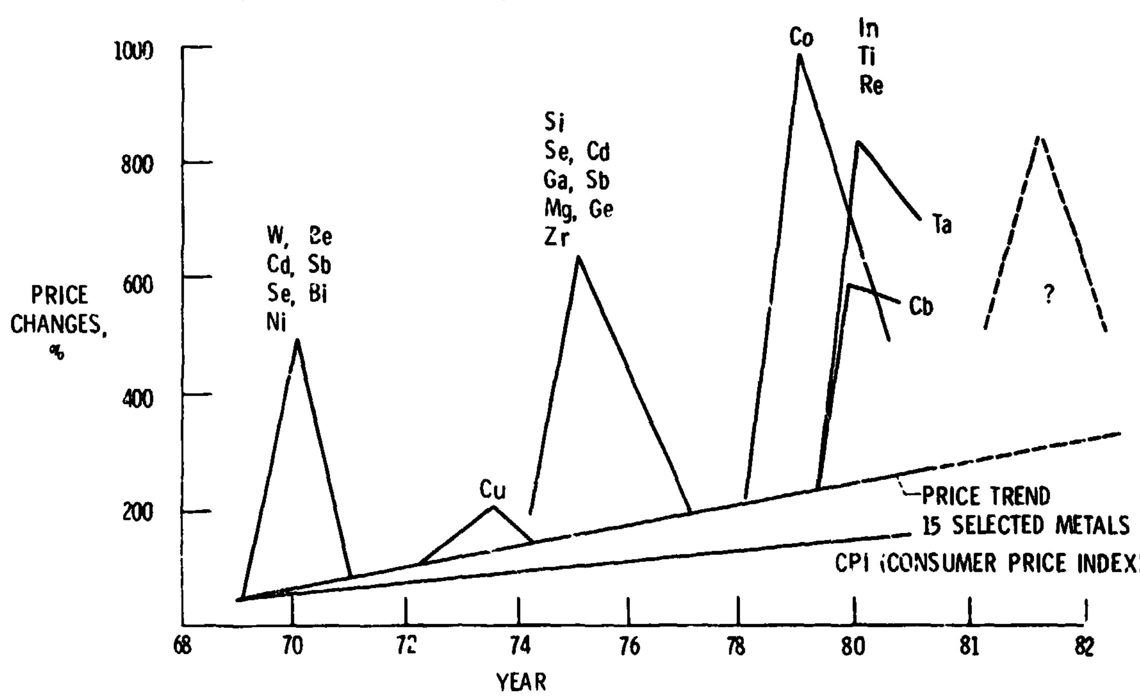
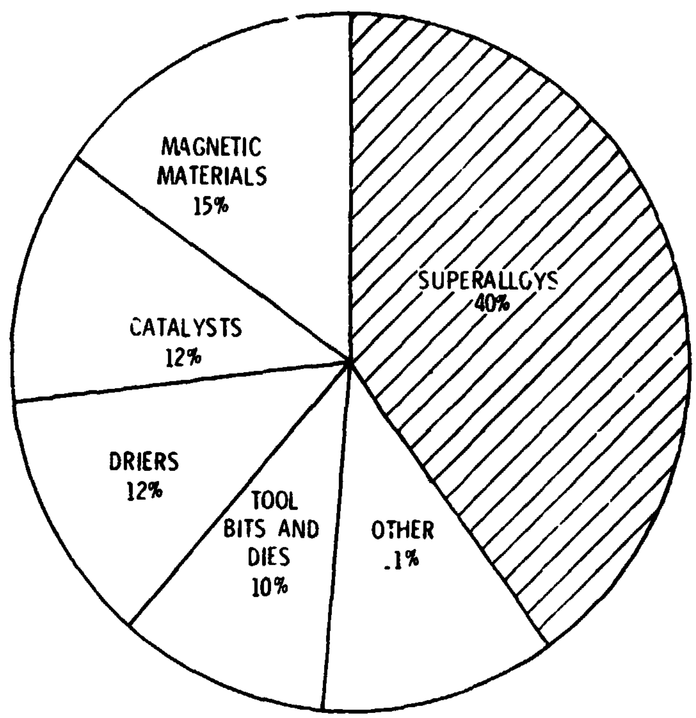
1981 SOURCES OF U. S. AEROSPACE STRATEGIC MATERIALS



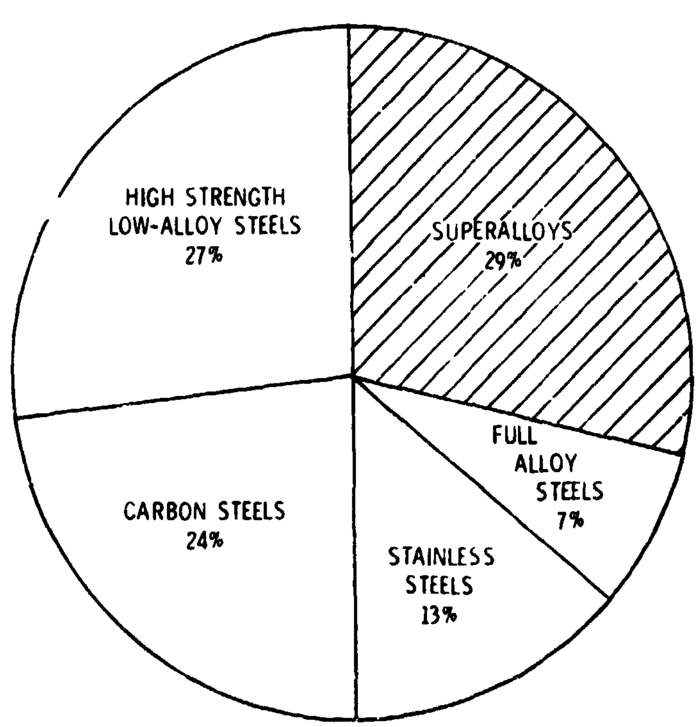
STRATEGIC MATERIAL PRICES ARE VOLATILE AND UNPREDICTABLE



DISTRIBUTION OF 1981 U. S. COBALT CONSUMPTION - 13.6 MILLION POUNDS



DISTRIBUTION OF 1980 U. S. COLUMBIUM CONSUMPTION - 6.5 MILLION POUNDS



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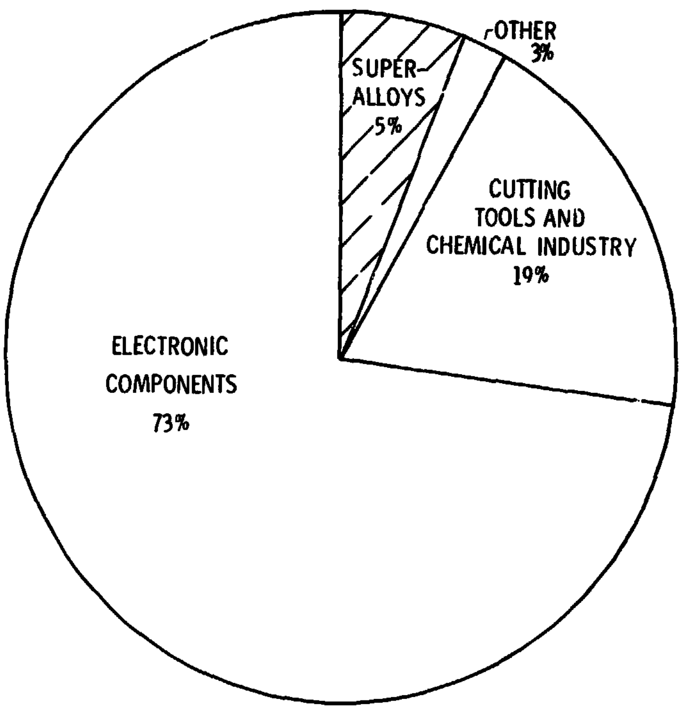
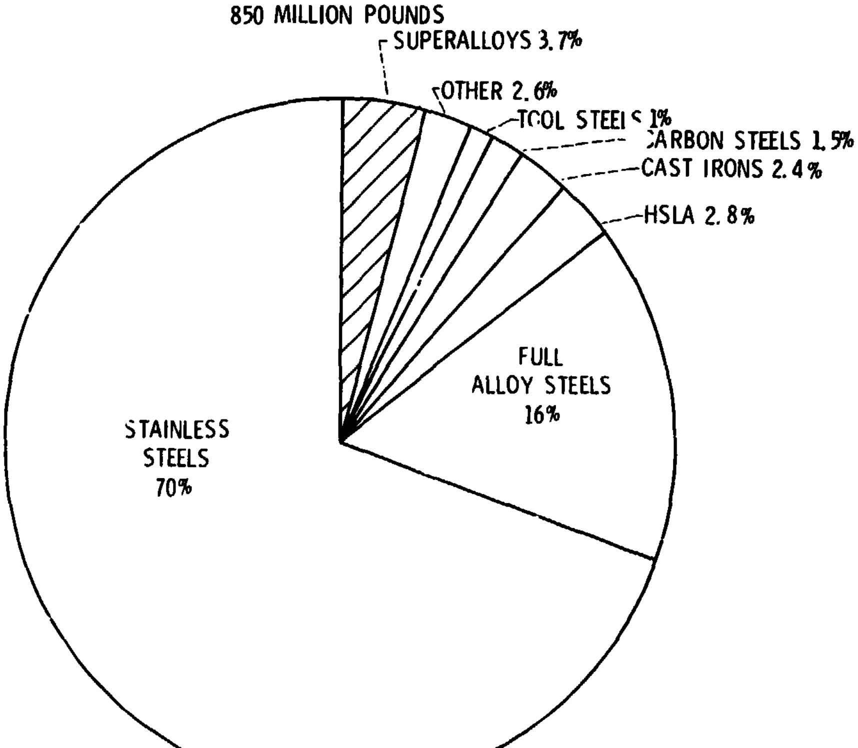
DISTRIBUTION OF 1981 U. S. TANTALUM CONSUMPTION - 1.3 MILLION POUNDS



DISTRIBUTION OF 1981 U. S. CHROMIUM CONSUMPTION

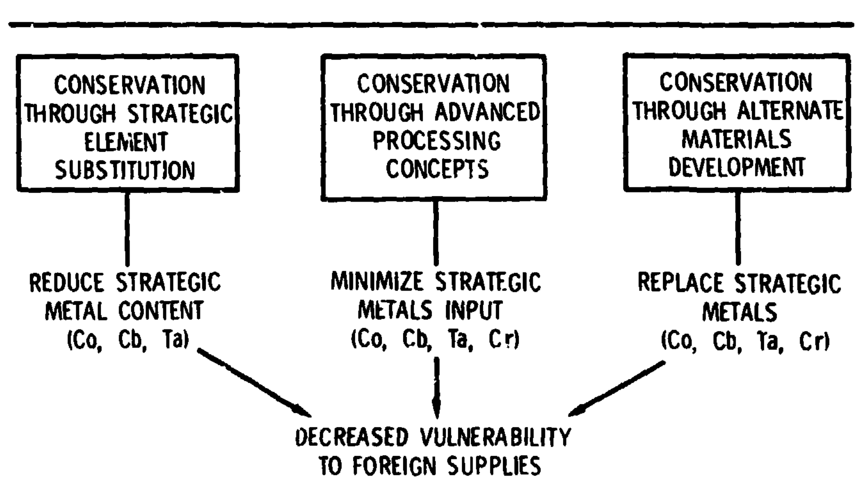
850 MILLION POUNDS



SUMMARY OF STRATEGIC METAL USE IN SUPERALLOYS

STRATEGIC ELEMENT	U.S. USE OF SUPERALLOYS %	MILLIONS OF lb
Co	40	5.44
Cb	29	1.89
Ta	5	0.07
Cr	3.7	31.45

CONSERVATION OF STRATEGIC AEROSPACE MATERIALS
(COSAM)



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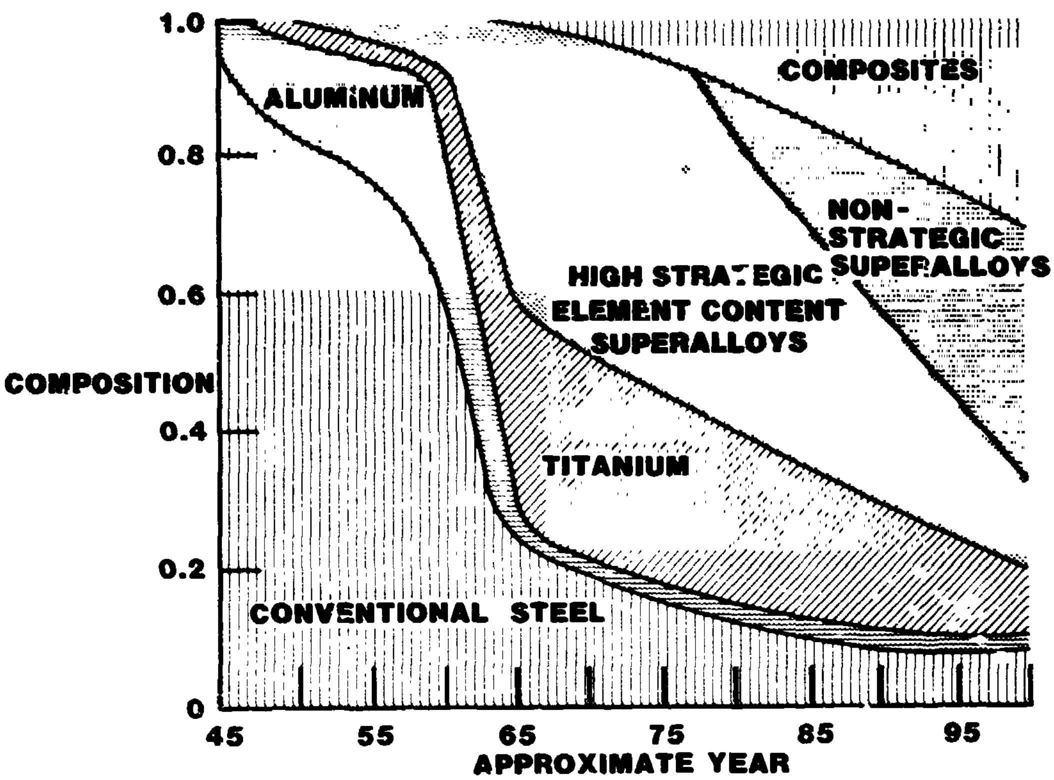
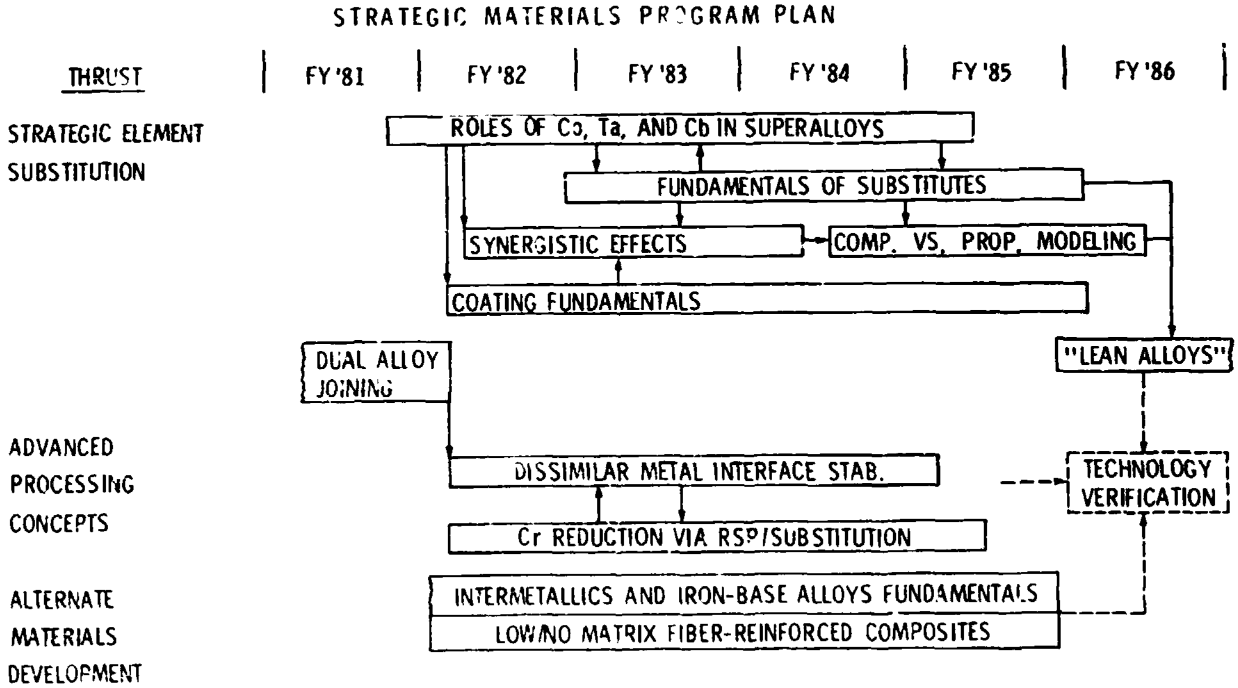
TRENDS IN AIRCRAFT GAS TURBINE MATERIAL USE



STRATEGIC MATERIALS PROGRAM PLAN



NASA LEWIS RESEARCH CENTER

- RESEARCH

–COMPOSITION/FABRICATION MODELING
–PHYSICAL/MECHANICAL PROPERTIES
–ENVIRONMENTAL/COATING EFFECTS
–PROPERTY/MICROSTRUCTURE CORRELATION

- PROGRAM MANAGEMENT

INDUSTRY	UNIVERSITIES
• RESEARCH	• RESEARCH
–FABRICABILITY	–MECHANICAL PROPERTIES
–PROCESSING	–MICROSTRUCTURE CHANGES
–PHYSICAL PROPERTIES	- MICROCHEMISTRY CHANGES
–MECHANICAL PROPERTIES	–ENVIRONMENTAL EFFECTS

COSAM PROGRAM SUMMARY

STRATEGIC ELEMENT SUBSTITUTION

PARTICIPATION	EFFORT
COBALT	
(1) COLUMBIA UNIVERSITY (2) PURDUE UNIVERSITY I SPECIAL METALS CORP. II BATTELLE COLUMBUS LABORATORIES NASA LEWIS RESEARCH CENTER	WASPALLOY, U-700, U-720, NIMONIC 115
NASA LEWIS RESEARCH CENTER	PM U-700
NASA LEWIS RESEARCH CENTER	RENÉ 150
COBALT/TANTALUM	
(3) CASE WESTERN RESERVE UNIVERSITY NASA LEWIS RESEARCH CENTER	Mar-M 247
TANTALUM	
(4) MICHIGAN TECHNOLOGICAL UNIVERSITY III GENERAL ELECTRIC R&D CEN. IV TRW INC. NASA LEWIS RESEARCH CENTER	Mar-M 247 B 1900 + HF
COLUMBIUM	
(5) CASE WESTERN RESERVE UNIVERSITY	IN 718

COSAM PROGRAM SUMMARY (CONCLUDED)

ADVANCED PROCESSING CONCEPTS

PARTICIPATION	EFFORT
CHROMIUM	
(6) UNIVERSITY OF ILLINOIS	RSP/Cr SUBSTITUTION - WASPALOY, IN 713 LC
ALL FOUR ELEMENTS	
NASA LEWIS RESEARCH CENTER	DUAL ALLOY INTERFACE STABILITY
COLUMBIUM	
NASA LEWIS RESEARCH CENTER V SPECIAL METALS CORP.	Sn EFFECTS IN IN 718

ALTERNATE MATERIALS DEVELOPMENT

PARTICIPATION	EFFORT
INTERMETALLIC COMPOUNDS	
NASA LEWIS RESEARCH CENTER (7) STANFORD UNIVERSITY	DEFORMATION MECH. FeAl, NiAl, CoAl
(8) DARTMOUTH COLLEGE	DUCTILITY NiAl
(9) TEXAS A&M UNIVERSITY	MODULI FeAl, NiAl, CoAl
(10) ILLINOIS INSTITUTE OF TECH. (ASM)	PHASE DIAGRAMS
IRON-BASE ALLOYS	
NASA LEWIS RESEARCH CENTER	LOW/NO Cr ALLOYS
(11) UNIVERSITY OF CONNECTICUT VI UNITED TECHNOLOGIES RESEARCH CENTER NASA LEWIS RESEARCH CENTER	IRON-BASE EUTECTICS
NASA LEWIS RESEARCH CENTER	IRON-BASE COMPOSITES
VII AIRESEARCH CASTING CO.	IRON-BASE ALLOYS
VIII UNITED TECHNOLOGIES RESEARCH CENTER	IRON-BASE ALLOYS

HIGHLIGHT SUMMARY

- DECREASING COBALT IN Ni-BASE SUPERALLOYS
  - –50% REDUCTION - MINOR EFFECT ON MECHANICAL PROPERTIES
  - –100% RED' 'TION - DECREASES RUPTURE LIFE/INCREASES CREEP RATE
    - IMPROVES OXIDATION/CORROSION RESISTANCE
    - CHANGES IN γ' %, CARBIDES, S.F. ENERGY, γ-γ' MISMATCH
- FINE GRAIN SIZE IN ALUMINIDES
  - –IMPROVES LOW TEMPERATURE DUCTILITY OF NiAl
  - –IMPROVES HIGH TEMPERATURE CREEP RESISTANCE OF FeAl

COSAM - FY '83 AND FY '84

- CONTINUE RESEARCH IN THREE THRUST AREAS OF
  - –STRATEGIC ELEMENT SUBSTITUTION
  - –ADVANCED PROCESSING CONCEPTS
  - –ALTERNATE MATERIALS DEVELOPMENT
- TECHNICAL TARGETS
  - –IDENTIFY EFFECTIVE SUBSTITUTES
  - –MODEL NON-STRATEGIC METAL ALLOYS
  - –UNDERSTAND DEFORMATION MECHANISMS OF ADVANCED MATERIALS
  - –DETERMINE ROLE OF PROCESSING ON LOW/NO STRATEGIC METAL ALLOYS AND INTERMETALLIC COMPOUNDS
- NEW INITIATIVE ("CRITICAL RESOURCES" NEW START) TARGETED FOR FY 85
  - –ADVOCACY BEGINS IN LATE 1982
  - –INDUSTRY INPUT/GUIDANCE SOUGHT
  - –INDUSTRY INTEREST REQUIRED FOR SUCCESS

N83 11284 D2-26

SUPERALLOY COMPOSITION MODELING*

Jeffrey Barefoot, Robert Jarrett, Juan Sanchez, and John Tien
Columbia University
New York, New York 10027

Superalloy design and re-design (element substitution) efforts can become less tedious and less costly if a predictive method can be developed to determine the γ/γ' phase fields, i.e. γ' volume fraction as a function of the multicomponent composition. In the past, the cluster variation method has been successfully used for binary alloys in which the precipitated phase is coherent with the matrix phase. We are extending this method for application to the multicomponent coherent γ'/γ nickel-base superalloys. It will be shown that the cluster variation method can accurately describe the equilibrium (incoherent) γ'/γ phase fields in the binary Ni-Al phase diagram. We have also computed the γ'/γ phase field, as a function of temperature, for the Ni-Cr-Al ternary phase diagram. A reasonable fit results between the calculated and the experimental diagrams. The modelling of the six component Ni-Cr-Al-Co-Mo-Ti base superalloy is also underway. The effect of Ni substitution for Co will be discussed.

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*Research supported by NASA under grant NASA NAG-3-57.

WHY MODEL THE γ/γ' PHASE FIELD

1.) ORGANIZE INTO AN EASILY COGNITIVE WHOLE WHAT IS KNOWN BY EXPERIMENT AND EXPERIENCE

2.) REDUCE THE NUMBER OF EXPERIMENTS NEEDED TO DEFINE THE EFFECTS OF A PARTICULAR ALLOYING ELEMENT

3.) ILLUMINATE THE CHANGES IN PARTITIONING THAT OCCUR AT HIGH TEMPERATURES

MODELING -- THE OPTIONS

1.) REGULAR OR SUBREGULAR SOLUTION MODELS

2.) EXPERIMENT THEN GEOMETRIC CONSTRUCTION USING THERMODYNAMIC EQUILIBRIUM CONSTRAINTS

3.) CLUSTER VARIATION METHOD

THE CLUSTER VARIATION METHOD

(TETRAHEDRON APPROX)

$$F = E - TS$$

$$E = \frac{W}{2}\sum_{ij} \varepsilon_{ij} Y_{ij}$$

$$S = k\left(6\sum_{ij} Y_{ij} \ln Y_{ij} - 5\sum_{ij} X_i \ln X_i - 2\sum_{ijkl} Z_{ijkl} \ln Z_{ijkl}\right)$$

ε_{ij} = PAIR INTERACTION PARAMETER (ENERGIES)

X_i, Y_{ij}, Z_{ijkl} = ARE POINT, PAIR, AND TETRAHEDRON PROBABILITIES

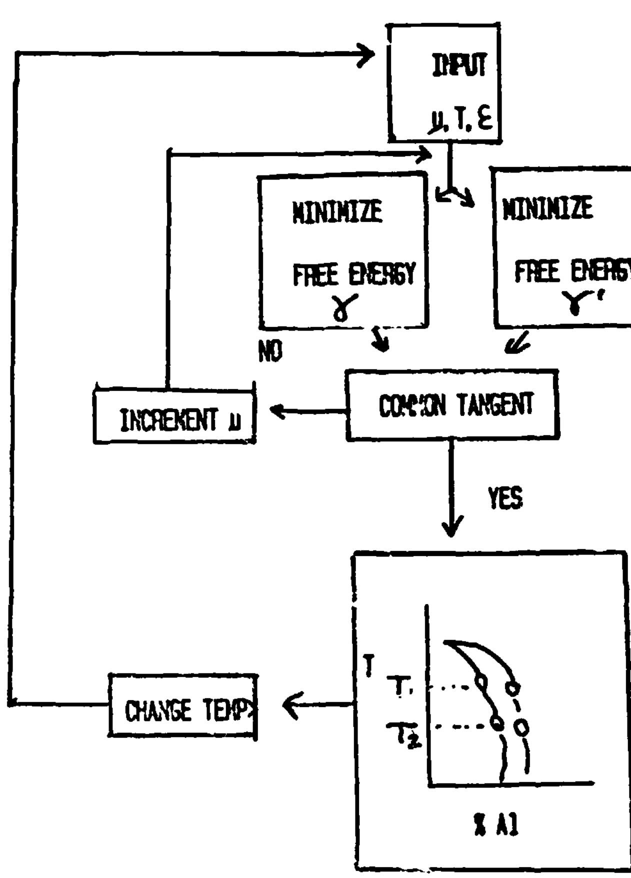
Ni-Al-Cr-Co-Mo-Ti SUPERALLOY STRATEGY

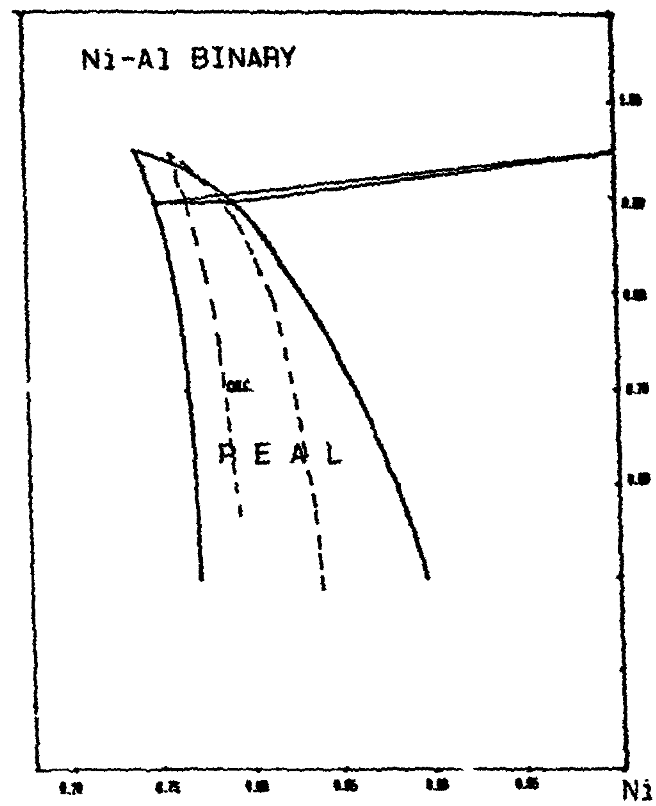
Ni-Al BINARY --> ε_{Ni-Al}

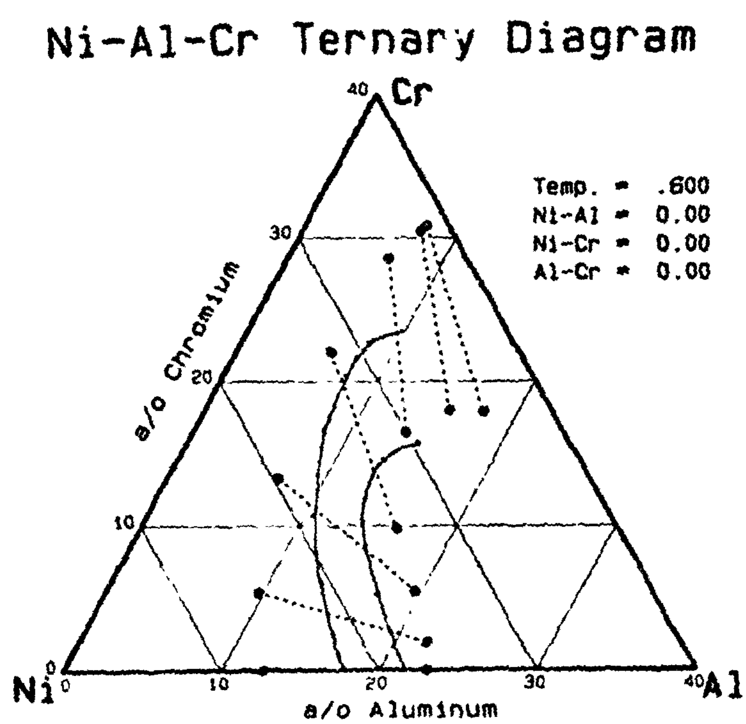
Ni-Al-Xx TERNARY'S --> ε_{Ni-Co} ε_{Al-Co} ε_{Ni-Cr} ε_{Al-Cr}
WHERE Xx = Co, Cr, Mo, Ti ε_{Ni-Mo} ε_{Al-Mo} ε_{Ni-Ti} ε_{Al-Ti}

Ni-Al-Xx-Yy QUATERNARY --> ε_{Co-Mo} ε_{Co-Cr} ε_{Co-Ti}
WHERE Xx ≠ Yy = Co, Mo, Ti, Cr ε_{Mo-Cr} ε_{Mo-Ti}
ε_{Cr-Ti}

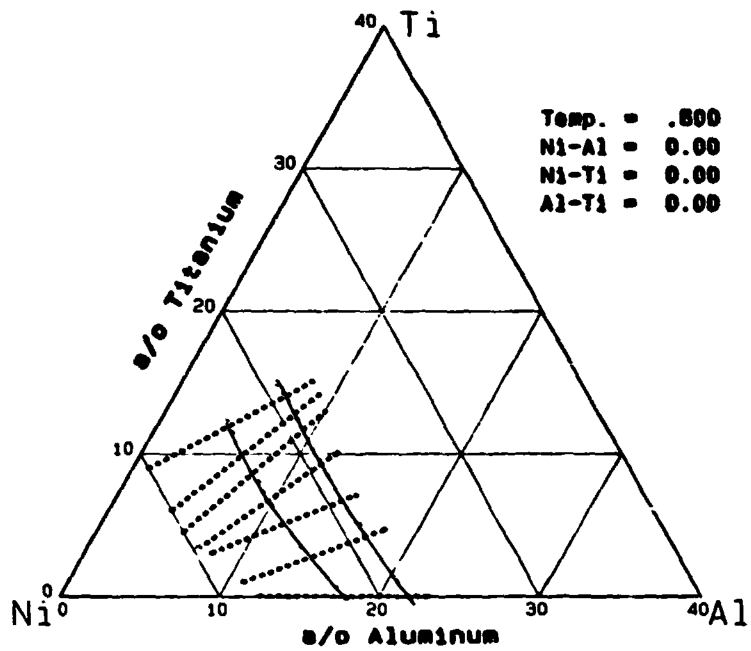
BASIC FLOWCHART FOR A SINGLE POINT ON THE PHASE DIAGRAM

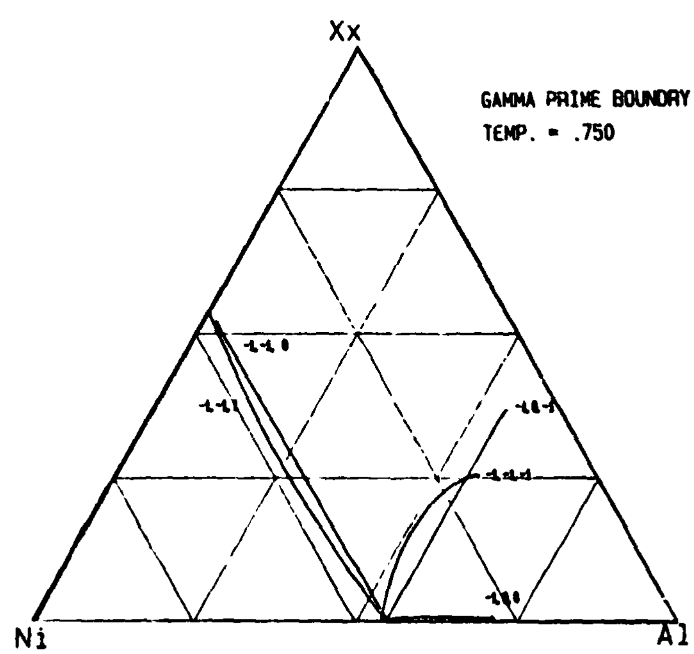


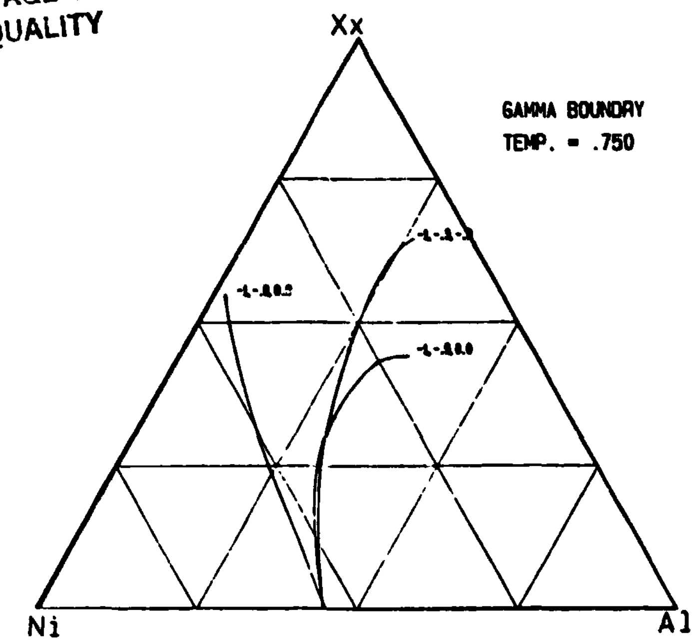

Ni-Al BINARY
REAL
Ni

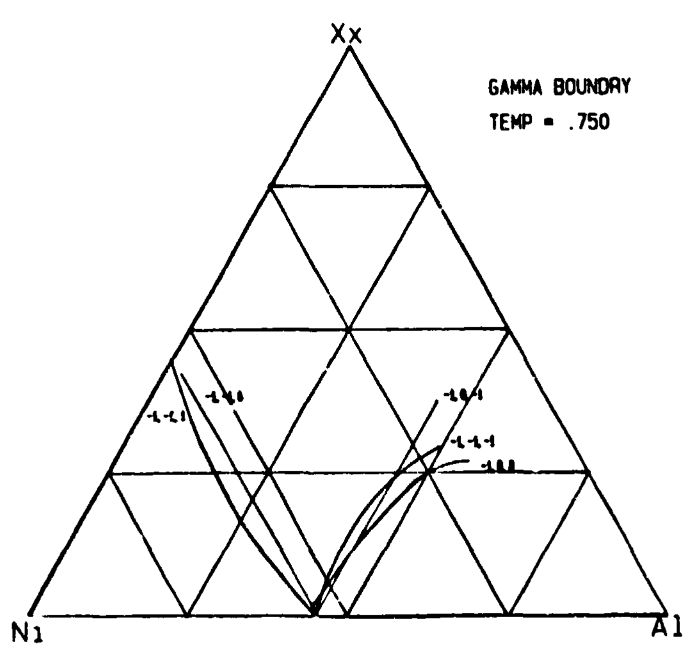

Ni-Al-Cr Ternary Diagram
Cr
Temp. = .600
Ni-Al = 0.00
Ni-Cr = 0.00
Al-Cr = 0.00
a/o Chromium
Ni
Al
a/o Aluminum
0
10
20
30
40

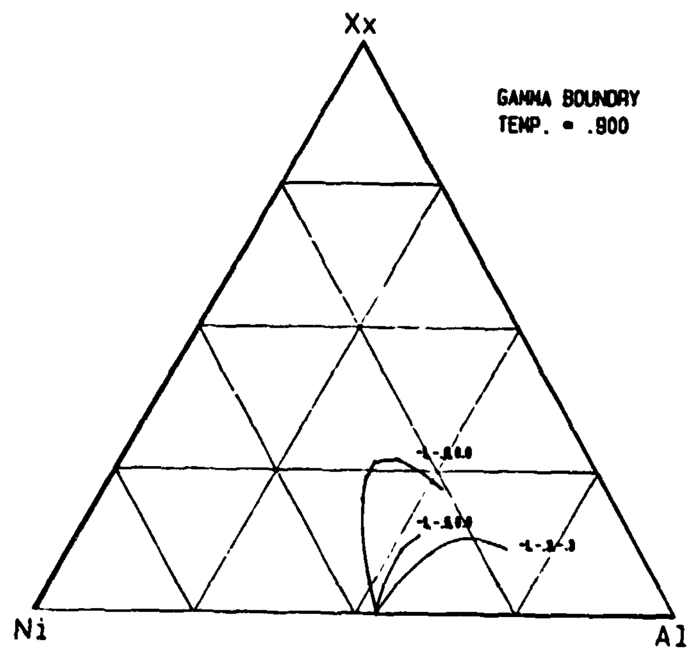
Ni-Al-Ti Ternary Diagram






Xx
GAMMA BOUNDRY
TEMP. = .750
Ni
Al


Xx
GAMMA BOUNDRY
TEMP = .750
Ni
Al



PROBLEMS TO BE SOLVED

1.) QUALITATIVE---> QUANTITATIVE BEHAVIOR

2.) DEVELOPE A MORE EXACT UNDERSTANDING OF INTERACTION ENERGIES, IN PARTICULAR THEIR DEPENDENCE ON LATTICE PARAMATERS AND CONCENTRATION

N83 11285 D3-26

PREPARATION OF LOW STRATEGIC METAL CONTENT SUPERALLOYS

F. E. Sczerzenie and G. E. Maurer
Special Metals Corporation
New Hartford, New York

Heats of modified NIMONIC 115 and UDIMET 720 were made with reduced levels of strategic element content (cobalt) to provide material for the Columbia University COSAM Research Program. Vacuum induction melted, and vacuum arc remelted ingots were hot rolled to 3/4-inch diameter bar. Hot workability was evaluated in terms of the ingot rolling behavior and the hot ductility of the as-rolled bar. Variations in workability and bar ductility were correlated to variations in incipient melting termperature and gamma prime solvus, both of which varied with cobalt content. Heat treatments were defined to yield, as far as possible, similar structures from alloy to alloy.

At the lowest cobalt levels N-115 workability was severely limited and the alloys could not be rolled to bar. Possible explanations for this behavior will be reviewed. DTA and metallographic data suggest that incipient melting in combination with heavy grain boundary carbide precipitation reduced ingot workability. Final heat treatment of modified alloys was complicated by the situation where the gamma prime solvus temperature was close to the incipient melting point. Under these conditions it may not be feasible to fully solution low Co alloys to obtain the large grain size required for optimum creep resistance.

FIGURE 1 - MELTING AND ROLLING OF LOW STRATEGIC METAL CONTENT SUPERALLOYS

VIM + VAR: 6" DIAMETER (125#)	
Alloy	Alloy
1 - N-115 with 14 Co (Standard)	5 - U-720 with 14.7 Co (Standard)
2 - N-115 with 10 Co	6 - U-720 with 7.5 Co
3 - N-115 with 5 Co	7 - U-720 with 0 Co
4 - N-115 with 0 Co	

Homogenize: N-115: 2200°F/24 Hours
U-720: 2135°F/24 Hours

Can, Roll 7" dia. to 3.9 RCS (60% Reduction)

Condition

Can, Roll 3.9" RCS to 1.6" RCS (83% Reduction)

Condition

Roll 1.6" RCS to 1.275" RCS (36% Reduction)

Roll 1.275" RCS to 0.75" Round (72% Reduction)

Cut Samples, Heat Treat and Test

FIGURE 2. H[illegible] [illegible] D N-115 4" RC[illegible]

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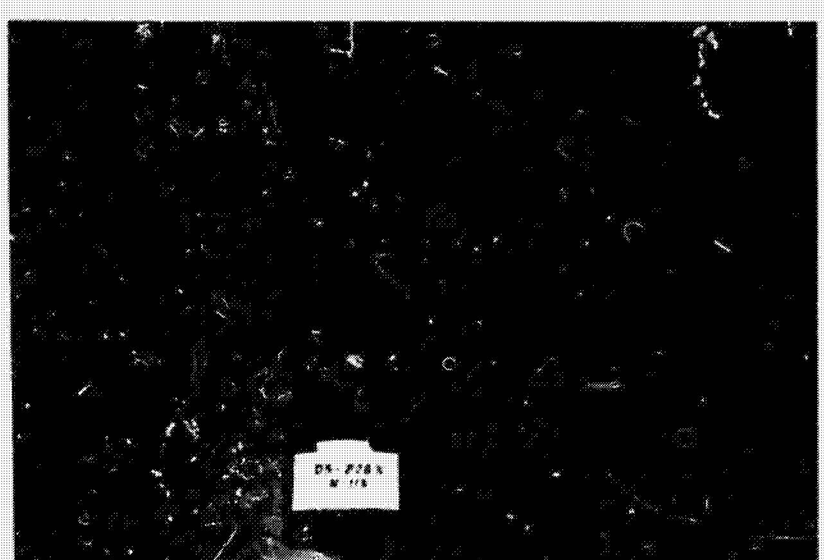

FIGURE 3 HEAT D5-2283 N-115 6" RCS

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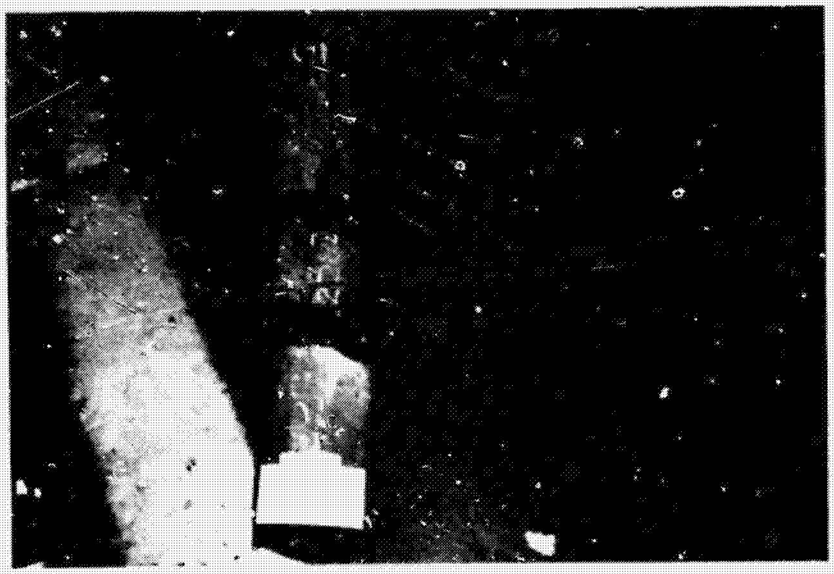

FIGURE 4 - HEAT D5-2282 N-115 6" RCS

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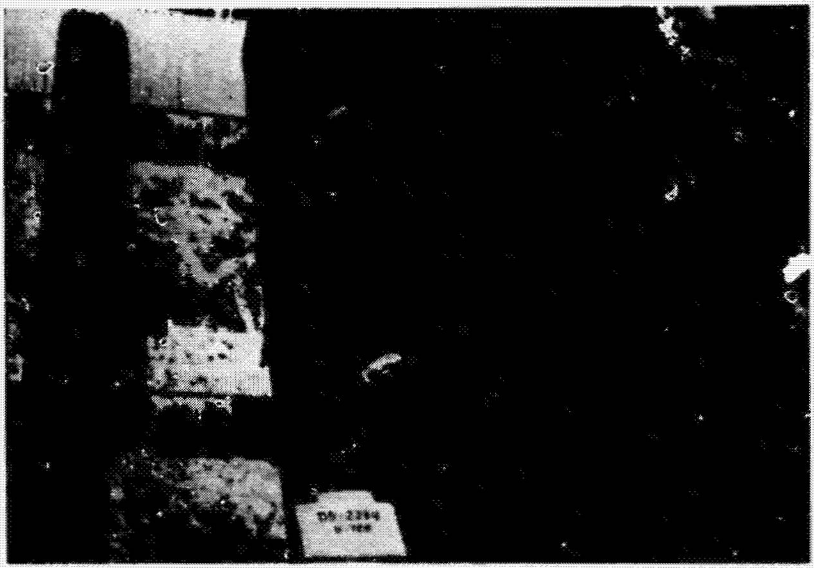

FIGURE 5 - HEAT D5-2284 U-720 4" RCS

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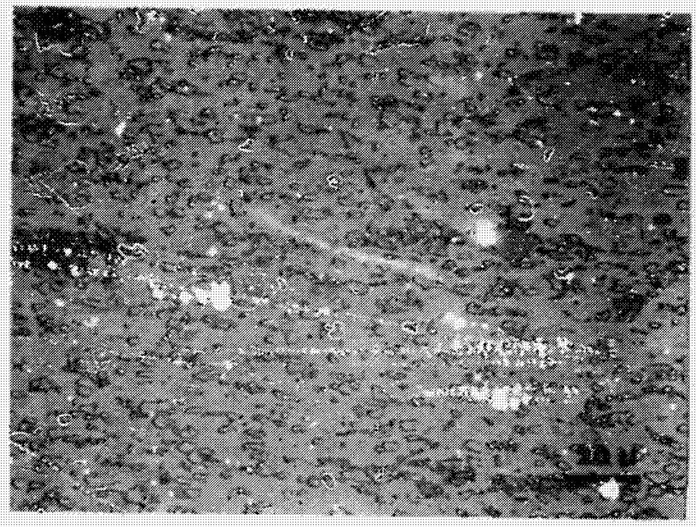
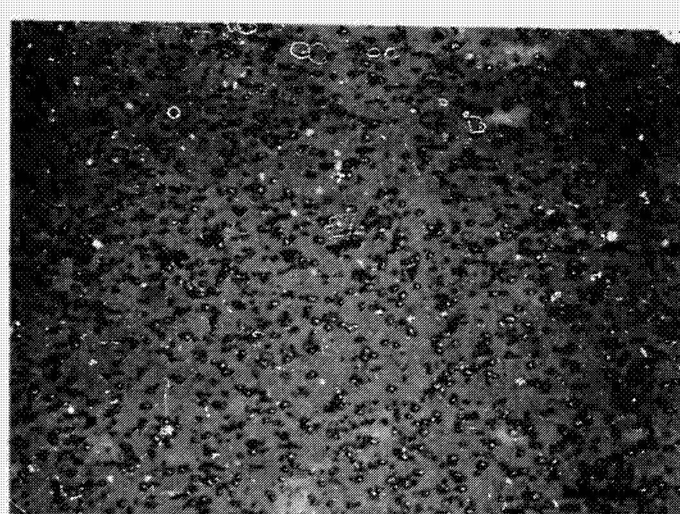


a



b

FIGURE 6 - As-Rolled Bar
a. D5-2280,
b. D5-2284.

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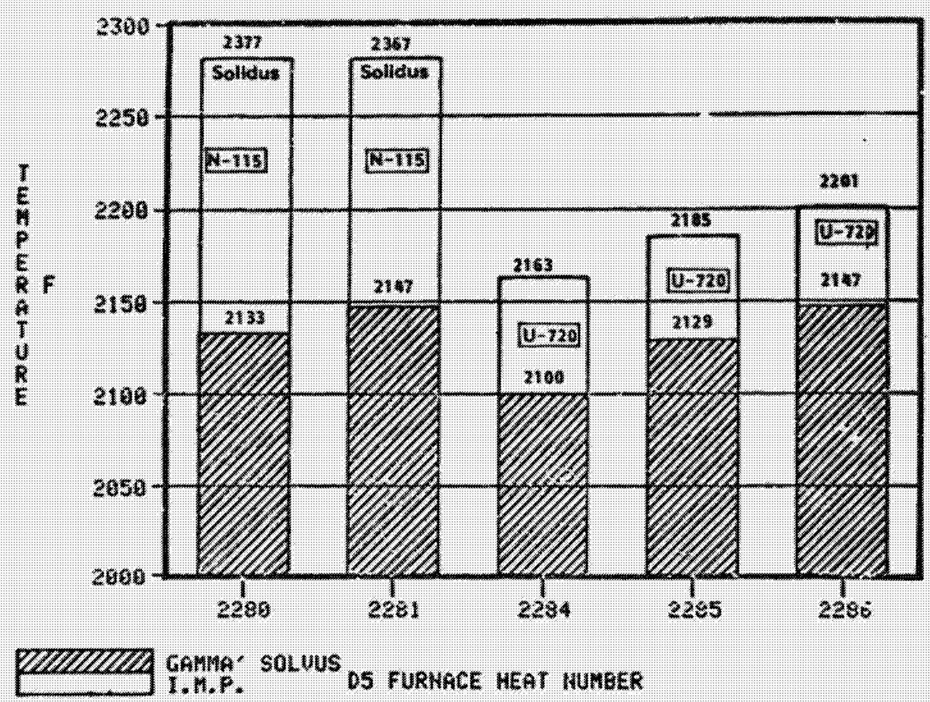

FIGURE 7 - DTA DATA, 3/4" DIA. AS ROLLED BAR
GAMMA PRIME SOLVUS AND INCIPIENT MELTING POINT

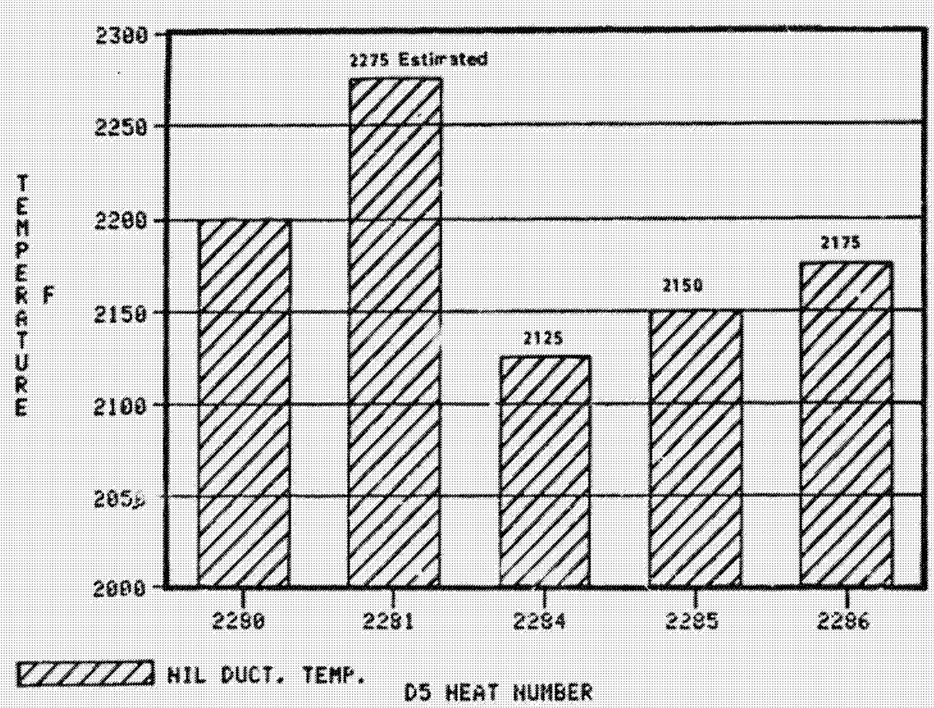

FIGURE 8 - NIL DUCTILITY TEMPERATURE OF 3/4" DIA. AS ROLLED BAR
TENSILE TEST, 2" PER SECOND

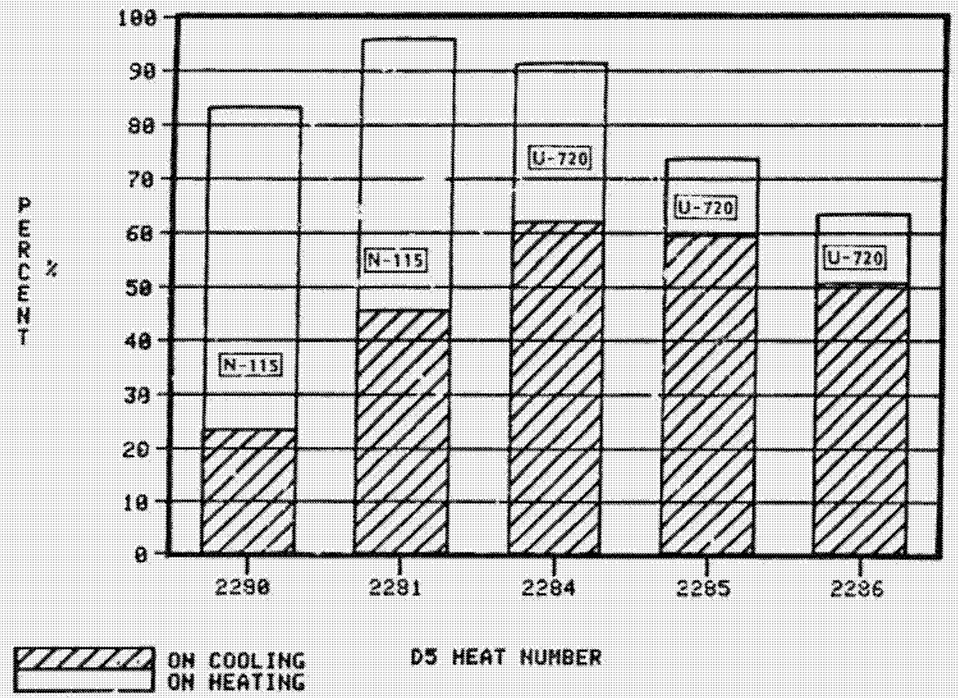

FIGURE 9 - 1900 F HOT DUCTILITY OF 3/4" DIA. AS ROLLED BAR
TENSILE REDUCTION OF AREA, 2" PER SECOND

FIGURE 10. D5-2280, 2175°F Solution Treated Microstructure. a. 100X, b. 500X.

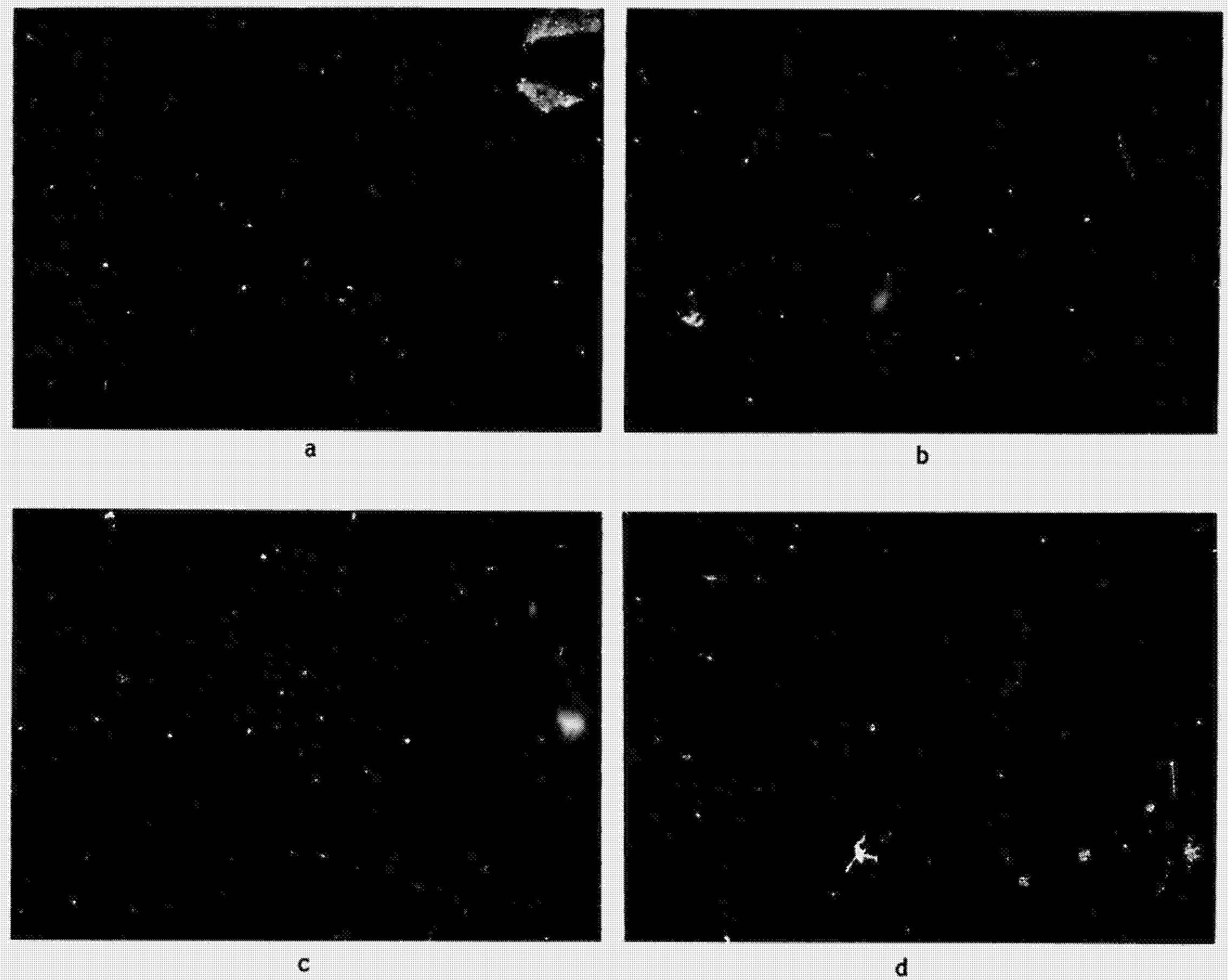

FIGURE 11. D5-2281, Heat Treated Microstructures; a and b 2175°F Solution Treated, c and d 2275°F Solution Treated.

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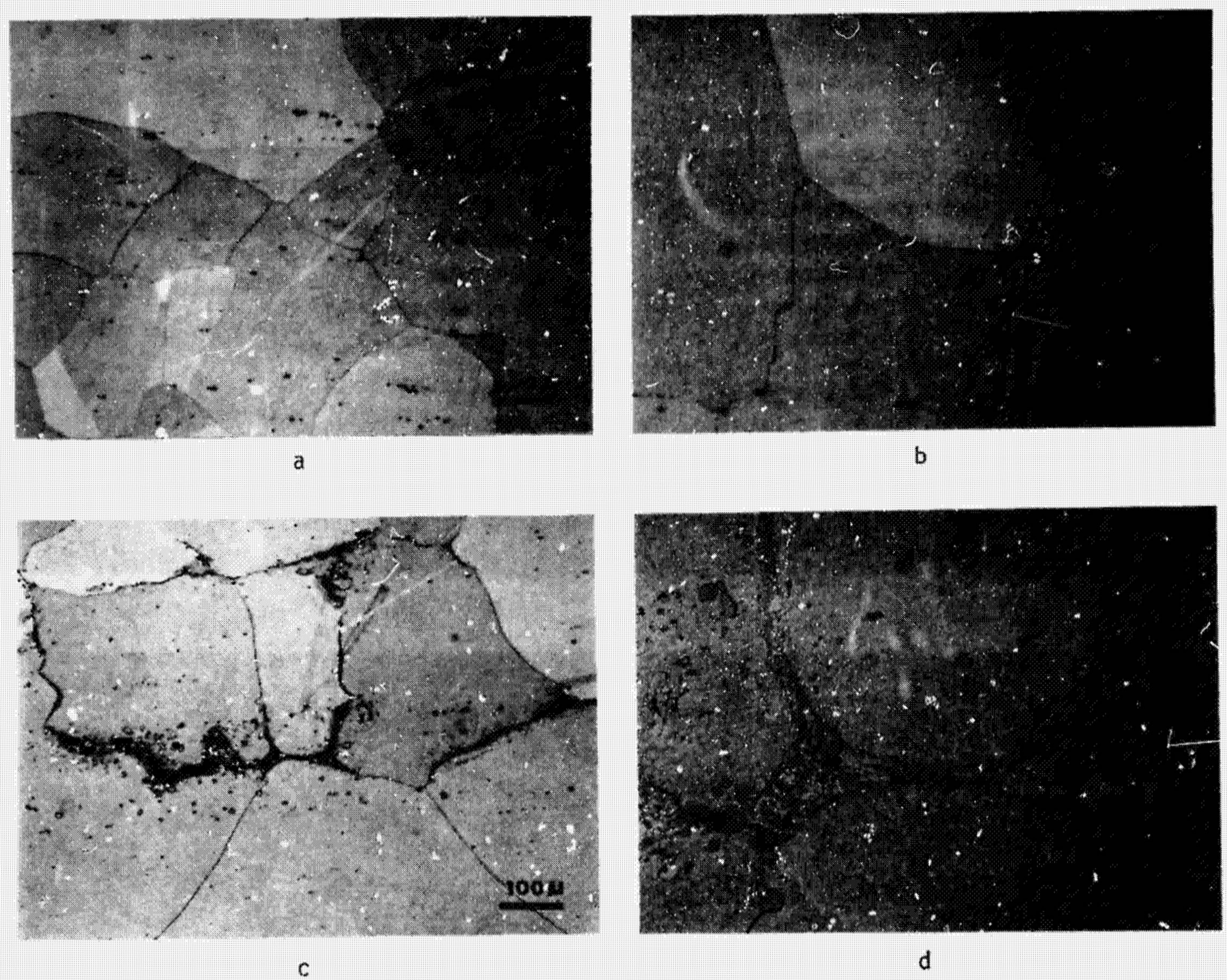

FIGURE 12. D5-2284 Heat Treated Microstructures;
a and b, 2135°F Solution Treatment,
c and d 2175°F Solution Treatment.

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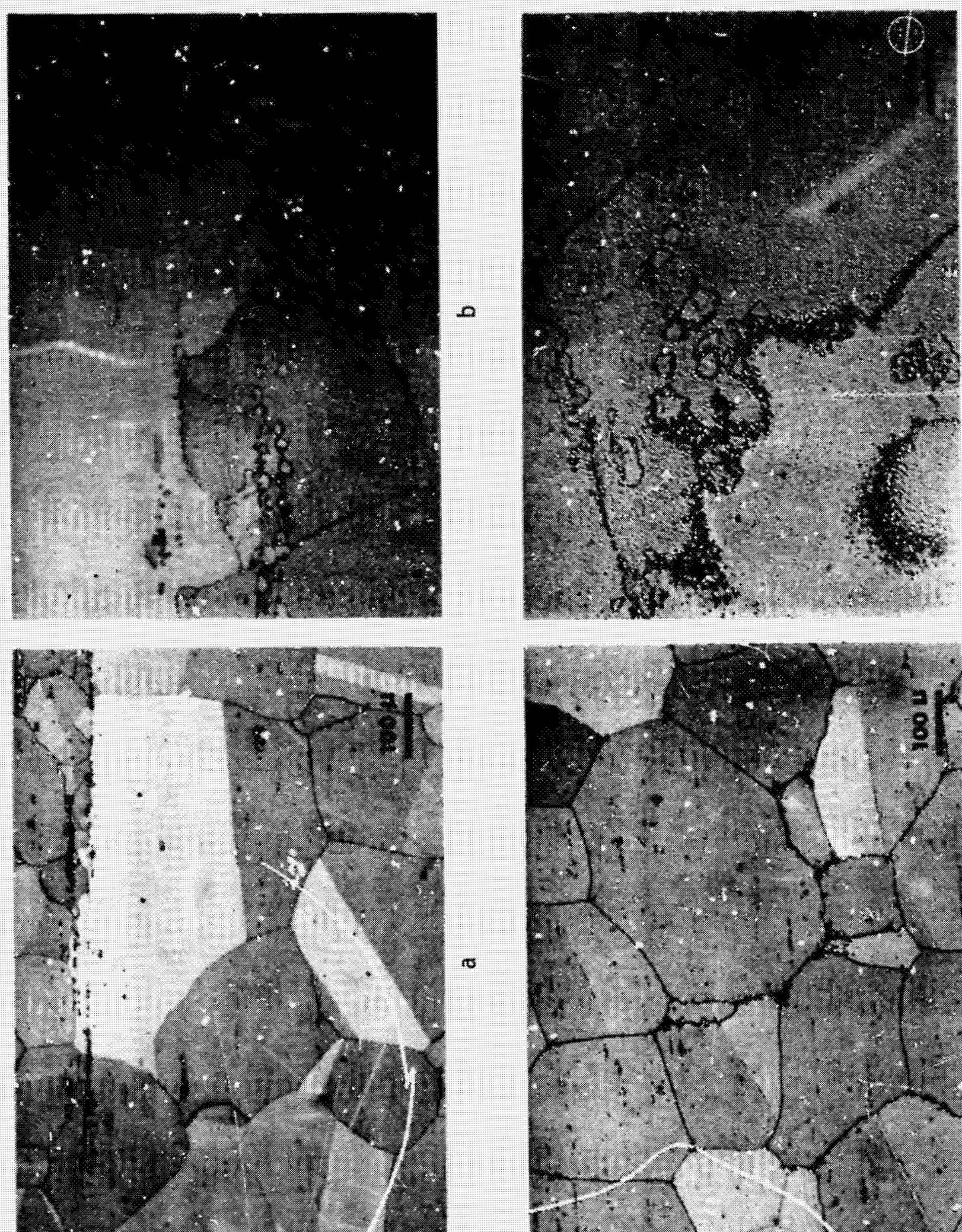

FIGURE 13. D5-2285 Heat Treated Microstructure;
a and b 2150°F Solution Treatment,
c and d 2175°F Solution Treatment.

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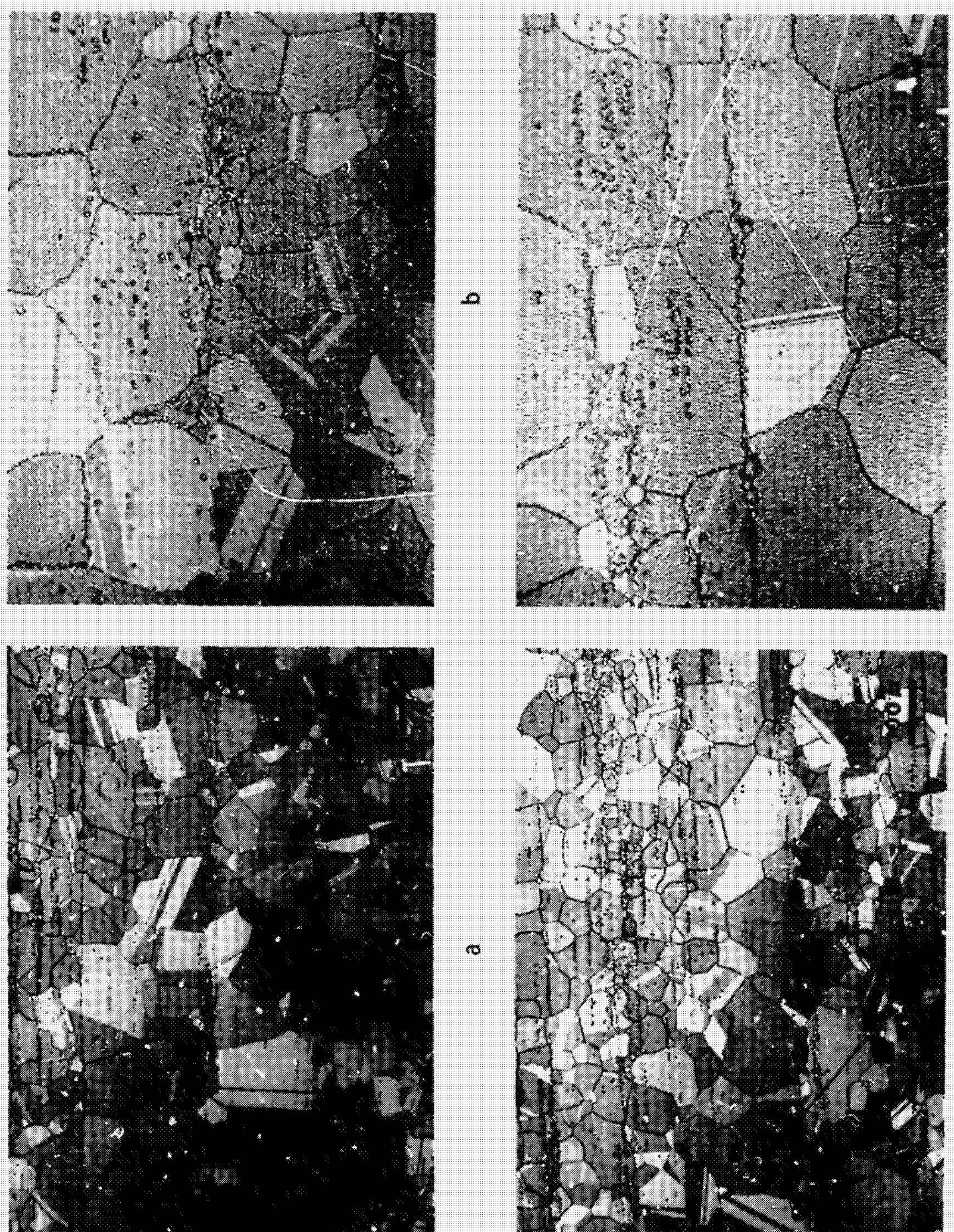

FIGURE 14. D5-2286 Heat Treated Microstructures;
a and b 2185°F Solution Treatment,
c and d 2200°F Solution Treatment.

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TABLE I

CHEMISTRIES OF LOW STRATEGIC METAL SUPERALLOYS

Alloy	Heat	Aim Chemistry	C	Cr	Co	Mo	W	Ti	Al	B	Zr	O ppm	N_2 ppm
N115	D5-2280	14 Co	.159	14.58	13.78	3.52	.01	3.95	4.91	.017	.001	9	6
N115	D5-2281	10 Co	.164	14.6	10.0	3.50	.01	3.87	4.75	.018	.003	12	8
N115	D5-2282	5 Co	.149	14.4	5.2	3.50	.01	3.97	4.88	.018	.003	6	6
N115	D5-2283	0 Co	.141	14.4	<0.15	3.45	.01	3.91	4.81	.018	.003	7	5
U-720	D5-2284	14.7 Co	.037	17.95	14.59	3.09	1.24	4.95	2.46	.031	.031	12	8
U-720	D5-2285	7.5 Co	.031	17.80	7.46	3.11	1.23	5.03	2.52	.031	.031	9	9
U-720	D5-2286	0 Co	.036	17.57	.01	3.04	1.23	4.99	2.48	.032	.030	8	10

TABLE II

DTA OF VAR INGOT

Heat No.	Alloy	% Cobalt	°F γ' Solvus	Incipient Melting	Point Solidus	Tangential Solidus	°F Liquidus
D5-2283	N115	0	2243	--	2308	2377	2470
D5-2282	N115	5	2221	--	2282	2367	2461
D5-2281	N115	10	2180	--	2282	2367	2462
D5-2280	N115	14 Control	2152	--	2282	2364	2462
D5-2286	U-720	0	2150	2215	2302	2357	2453
D5-2285	U-720	7.5	2132	2205	2305	2343	2457
D5-2284	U-720	14.7 Control	2110	2188	2313	2340	2462

TABLE III

DTA OF 3/4" DIA. BAR, AS ROLLED

Heat No.	Alloy	Aim Co Content %	On Heating γ' Solvus °F	2nd γ' Solvus °F	3rd γ' Solvus °F	Incipient Melting Temp. °F
D5-2280	N115	14	2134	2133	2133	ND[1]
D5-2281	N115	10	2152	2147	2147	ND[1]
D5-2284	U-720	14.7	2120	2097	2100	2163
D5-2285	U-720	7.5	2131	2127	2129	2185
D5-2286	U-720	0	2167	2145	2147	2201

[1]ND = None Detected

TABLE IV - HOT TENSILE PROPERTIES OF N-115, 3/4 INCH DIAMETER BAR, AS ROLLED[1]

HEAT NO.	TEST TEMP. (°F)	UTS KSI	0.2% YS KSI	% El	% R.A.
D5-2281	2250	13.6	13.6	12.6	12.6
D5-2280	2200	25.1	25.1	3.5	1.8
D5-2281	2200	24.6	24.6	71.0	56.5
D5-2280	2150	26.5	26.5	118.6	73.4
D5-2281	2150	28.5	28.5	138.2	79.4
D5-2280	2100	31.0	30.7	88.9	87.3
D5-2280	2100	30.7	30.7	110.8	89.3
D5-2281	2100	29.1	29.1	142.2	92.2
D5-2281	2100	30.5	30.5	102.3	90.3
D5-2280	2050	35.6	35.1	91.3	93.2
D5-2281	2050	34.2	33.4	128.4	95.3
D5-2280	2000	40.8	39.1	83.3	95.2
D5-2281	2000	42.4	40.7	85.4	95.4
D5-2280	1950	53.1	50.6	66.4	95.5
D5-2280	1950	50.6	48.7	77.2	95.5
D5-2281	1950	53.5	50.2	73.8	96.0
D5-2281	1950	54.5	50.3	93.3	96.4
D5-2280	1900	60.0	57.4	73.5	95.2
D5-2281	1900	46.9	44.4	101.0	96.5
D5-2280	1800	101.7	91.4	48.9	83.2
D5-2281	1800	98.4	84.7	62.5	96.1
D5-2280	1800[2]	105.6	95.0	32.5	23.5
D5-2281	1800[2]	99.4	90.2	43.5	45.5

[1]Constant stroke rate, 2 inches per second.

[2]On-cooling from 2100°F.

TABLE V - HOT TENSILE PROPERTIES OF U-720, 3/4 INCH DIAMETER BAR, AS ROLLED[1]

HEAT NO.	TEMP. (°F)	UTS KSI	0.2% YS KSI	% El	% R.A.
D5-2284	1800	95.2	88.7	58.7	91.4
D5-2285	1800	93.5	88.1	54.5	73.1
D5-2286	1800	84.0	77.9	52.8	63.5
D5-2284	1900	65.8	64.4	68.5	94.9
D5-2285	1900	63.8	61.5	65.8	91.4
D5-2286	1900	59.9	58.3	71.8	88.1
D5-2284	2050	36.1	35.9	102.0	87.9
D5-2285	2050	28.7	28.7	127.1	90.5
D5-2286	2050	37.1	36.4	95.1	93.0
D5-2284	2100	34.8	34.8	91.9	75.6
D5-2285	2100	32.7	32.7	150.4	68.9[2]
D5-2286	2100	31.6	31.4	89.4	85.2
D5-2284	2025 -> 1800	100.2	94.1	51.2	62.1
D5-2285	2025 -> 1800	97.8	--	43.9	59.5
D5-2286	2025 -> 1800	94.9	88.4	44.2	50.6
D5-2284	2125	15.1	15.1	0	1.2
D5-2285	2125	31.8	31.8	89.2	56.6
D5-2286	2125	30.8	30.8	94.6	68.0
D5-2285	2150	12.7	12.7	0	1.1
D5-2286	2150	25.6	25.6	81.5	55.5
D5-2286	2175	14.7	14.7	0	1.1

[1]Constant stroke rate 2 inches per second.

[2]Measurements at end of actuator movement; specimen did not break.

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TABLE VI - HEAT TREATMENT STUDY FOR LOW STRATEGIC METAL SUPERALLOYS

Heat No.	Solution Temp.	Grain Size ASTM No. Primary	Secondary	Comment
2280	2175	3 to 4	30% 0 to 1	Secondary in random bands.
	2200	00 to 1	70% 4 to 5	Secondary in random bands.
2281	2175	2 to 4	30% 00 to 1	Secondary in random patches, lamellar gamma prime.
	2200	1 to 2	--	Lamellar gamma prime; possible incipient melting.
	2225	--	--	Heavy lamellar gamma prime and nodular gamma prime at grain boundaries.
2284	2135	00 to 2		Random variation of grain size.
	2150	00 to 2		" " " "
	2175	000 to 1	--	Incipient melting.
2285	2135	4 to 6	00 to 1	50% coarse grained toward center of bar.
	2150	0 to 2	30% 4 to 5	Secondary grain size in bands.
	2175	--	--	Incipient melting.
2286	2135	8 to 9		
	2150	3 to 6	40% 8 to 9	Secondary grain size in stringers
	2175	2 to 5	40% 6 to 7	No incipient [illegible]

TABLE VII - HEAT TREATMENTS FOR LOW STRATEGIC METAL CONTENT SUPERALLOYS

Heat Number	Treatment
D5-2280 and D5-2281	Solution 2175°F, 4 Hours, Furnace Cool to 1832°F, Air Cool
D5-2284	Solution 2135°F, 4 Hours, Air Cool Partial Solution 1975°F, 4 Hours, Air Cool Age (1550°F, 24 Hours, Air Cool (1400°F, 16 Hours, Air Cool
D5-2285	Solution 2150°F, 4 Hours, Air Cool Plus Partial Solution and Age, Same as D5-2284
D5-2286	2175°F, 4 Hours, Air Cool Plus Partial Solution and Age, Same as D5-2284

N83 11286 D4

ROLE OF COBALT IN NICKEL BASE SUPERALLOYS*

Robert Jarrett, Jeffrey Barefoot, John Tien, and Juan Sanchez
Columbia University
New York, New York

We report on the progress of the research program aimed at understanding the role of cobalt in nickel-base superalloys. The three systems discussed, Waspaloy, Udimet 700 and Nimonic 115, are representative of Ni-Cr-Co-Al-Ti-Mo superalloys strengthened by a heavily alloyed matrix, coherent γ' precipitates and carbides at the grain boundaries. These alloys differ in the amount of γ' -- Waspaloy with $\sim$ 20%, Udimet 700 with $\sim$ 45% and Nimonic 115 with $\sim$ 55% γ'. Accordingly the way cobalt (or substituting for cobalt) affects the γ' solvus temperature and the chemical partitioning in each alloy is different. Using the results obtained for the three systems a generalized understanding of the role of cobalt is discussed. Microstructure and in-situ and extracted phase STEM micro-analysis results will be used to explain cobalts effect on mechanical properties.

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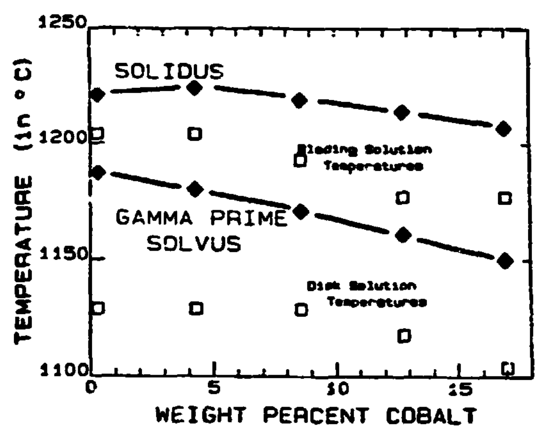
*Research supported by NASA under grant NASA NAG 3-57.

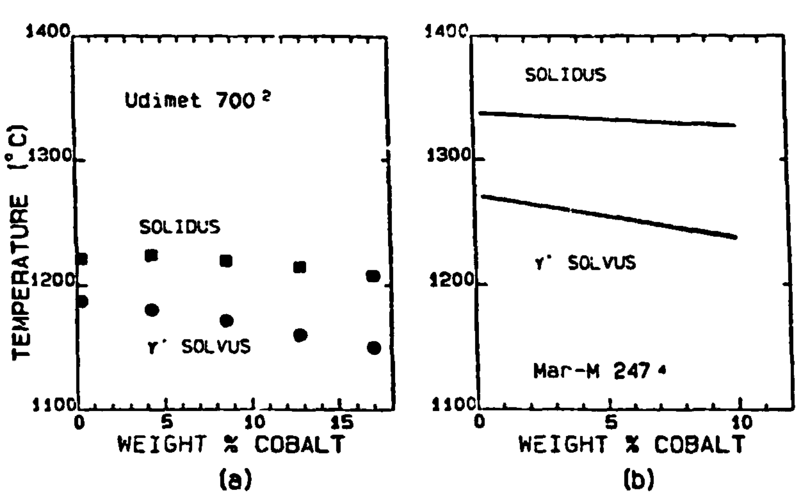
Nominal Compositions of Sev .1 Nickel-Base Superalloys*

	% γ'		Ni	Co	Cr	Al	Ti	Mo	W	Ta	Hf	B	Zr	C
Waspaloy[4]	20	w/o	58	13.5	19.5	1.3	3.0	4.3				.006	.06	.08
		a/o	56	13.0	21.4	2.7	3.6	2.6				.03	.04	.38
Udimet 700[2]	45	w/o	53	18.5	15.0	4.3	3.5	5.2				.030		.08
		a/o	50	17.4	16.0	8.8	4.1	3.0				.15		.37
Mar-M247[3]	55	w/o	60	10.0	8.2	5.5	1.0	0.6	10.	3.	1.5	.020	.09	.16
		a/o	61	10.1	9.2	12.2	1.2	0.4	3.	1.	0.5	.11	.06	.79

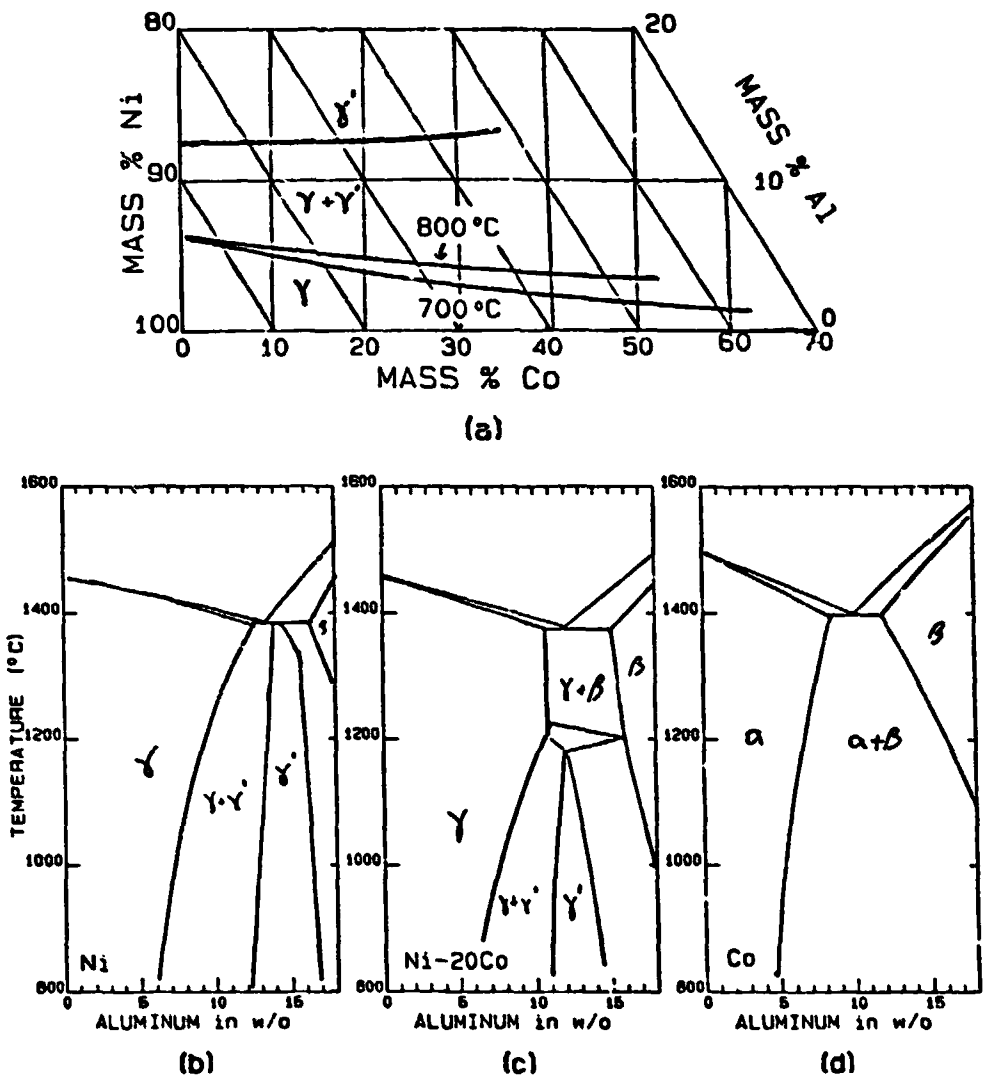
*From the International Nickel Company, Inc. Handbook on "High Temperature, High Strength Nickel-Base Alloys," 1979.



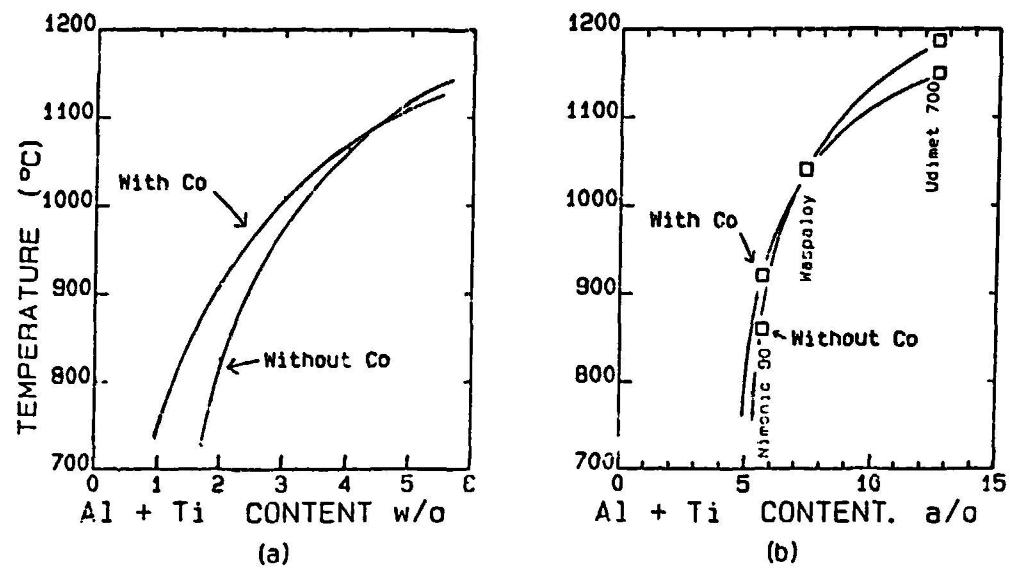
DTA Results and Solution Temperatures
for the Heat Treatments



γ' solvus and solidus temperatures versus cobalt content for (a) Udimet 700[2] and (b) Mar-M247.[3]

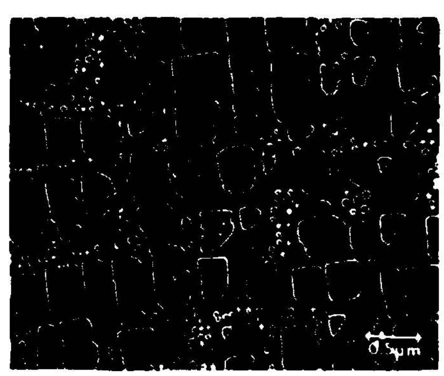


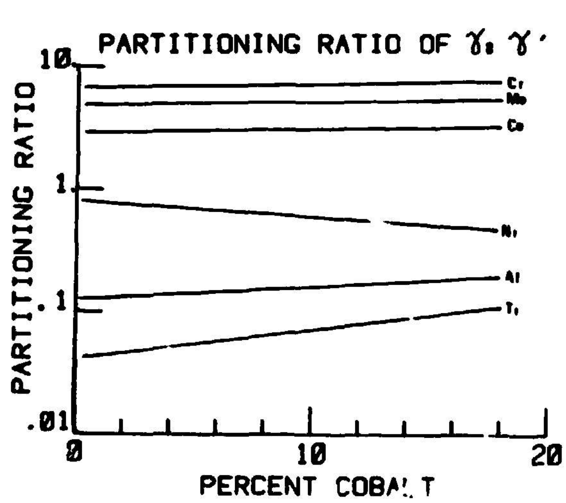
The Ni-Co-Al ternary phase diagram. Fig. 2a shows a composite of isothermal sections at $700^{\circ}C$ and $800^{\circ}C$ from Davies et al.[11] Fig. 2b is the Ni-Al binary, Fig. 2c is the 80Ni-20Co compositional section, and Fig. 2d is the Co-Al binary.[10]

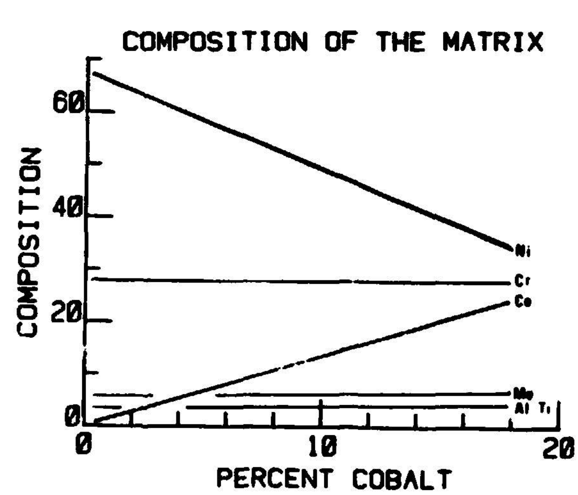


Pseudo-binary phase diagrams. Fig. 3a: Heslop's[9] pseudo-binary of the Ni-20Cr and *Ni-20Cr-20Co matrices. Fig. 3b: Composite from* γ' *solvus results.*[2,4,9] In both diagrams note that cobalt decreases the solubility of (Al+Ti) and in the higher γ' volume fraction alloys cobalt decreases the γ' solvus temperature.

FINE CUBIC γ' SURROUNDED BY THE γ MATRIX







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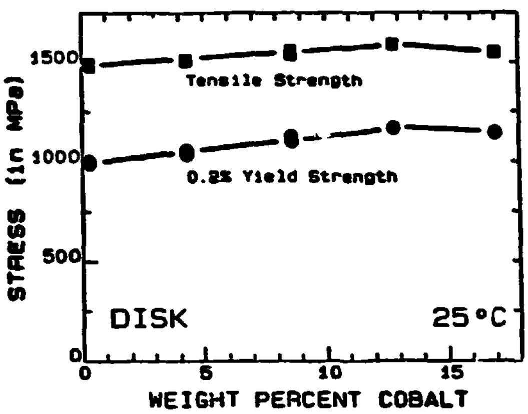
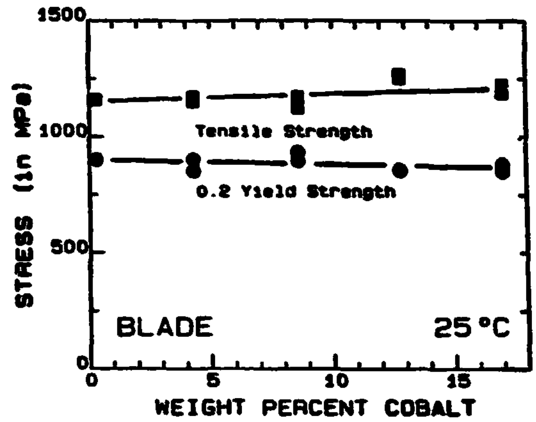


Room Temperature Tensile and Yield Strengths of Disk UDIMET 700



Room Temperature Tensile and Yield Strengths of Blade UDIMET 700

Yield Strength Parameters of Alloys After Disk Heat Treatments Using Equation 2 (25)

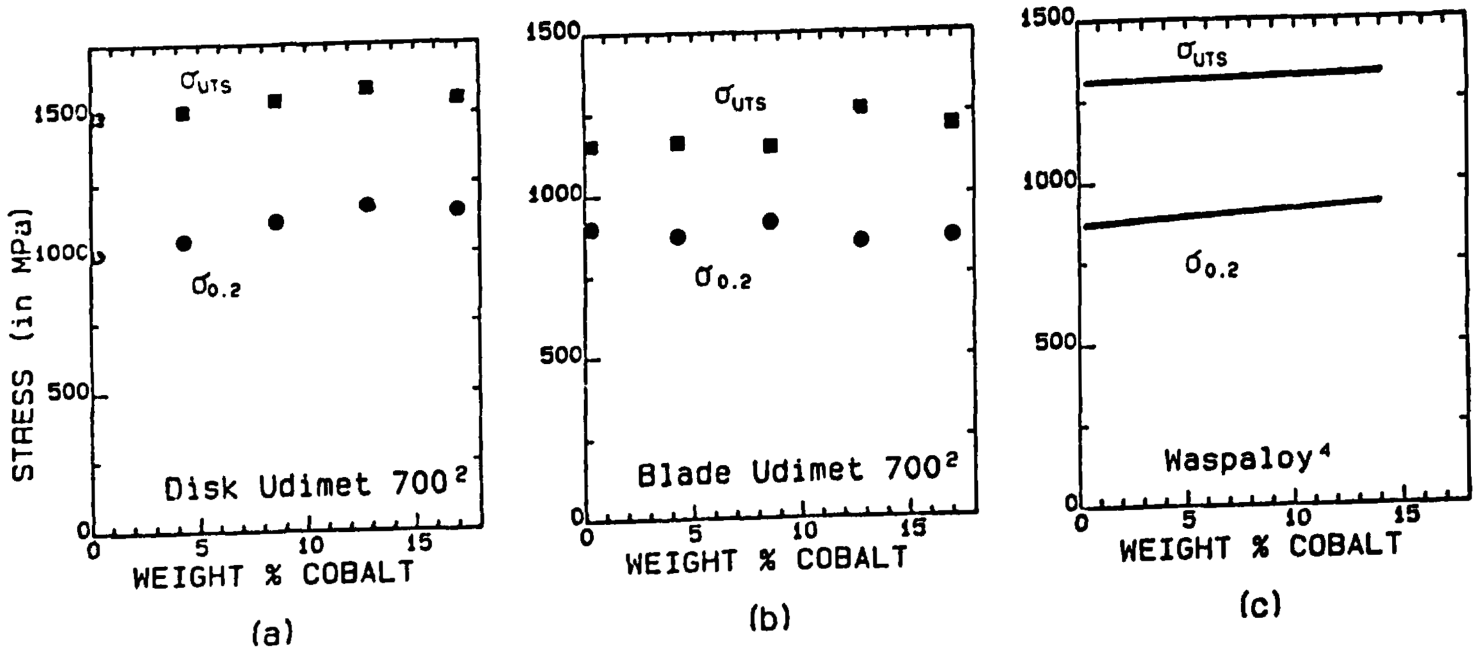
Cobalt Content (w/o)	Volume Fraction Fine γ'	Observed Yield Strength	Calculated* Yield Strength	Calculated Γ_{APB}
0.0	.288	997 MPa	1031 MPa	161 mJ/m^2
4.3	.319	1046	1080	162
8.6	.342	1113	1114	165
12.8	.367	1169	1150	167
17.0	.364	1146	1146	165

*Using Eq. 2 with constant Γ_{APB} of 165 mJ/m^2

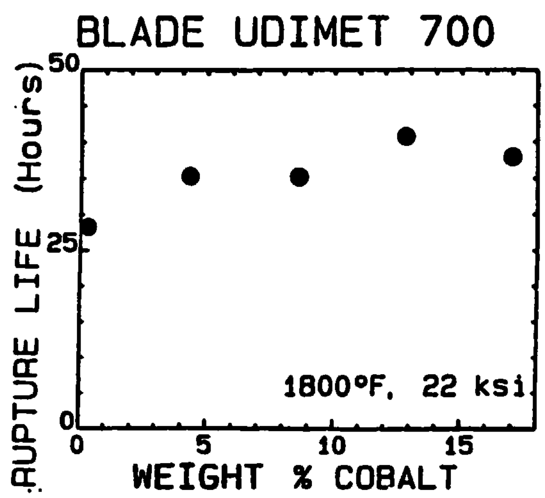
$$\Delta\tau_Y = (\Gamma_{APB}/2b)((4\Gamma_{APB} r_o f/\pi\phi)^{1/2} - f)$$

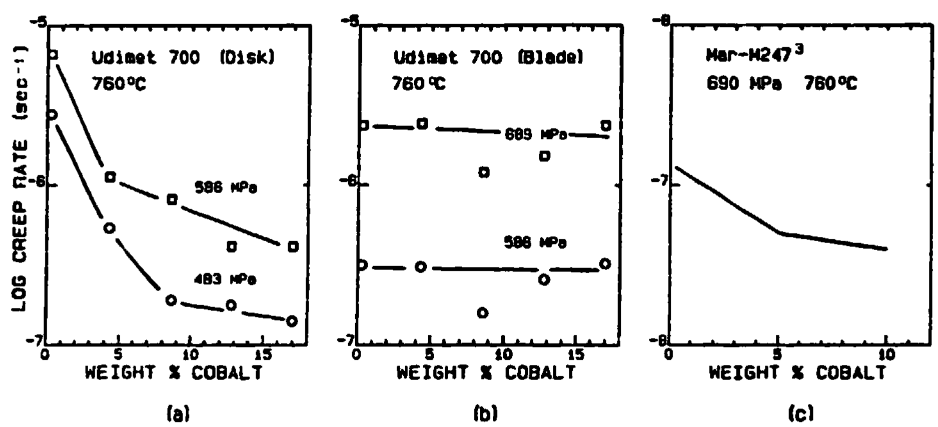
Partitioning Ratios (Wt Pct in γ/Wt Pct in γ')

Alloy Cobalt Content	Co Disk	Co Blade	Ni Disk	Ni Blade	Cr Disk	Cr Blade	Al Disk	Al Blade	Ti Disk	Ti Blade	Mo Disk	Mo Blade
17.0	2.92	2.57	0.56	0.65	6.3	7.3	0.25	0.34	0.13	0.26	4.2	2.2
8.6	3.08	2.54	0.68	0.73	6.2	5.1	0.22	0.36	0.09	0.26	4.2	2.2
< 0.1	—	—	0.83	0.84	6.1	3.9	0.19	0.38	< 0.05	0.26	4.2	2.2

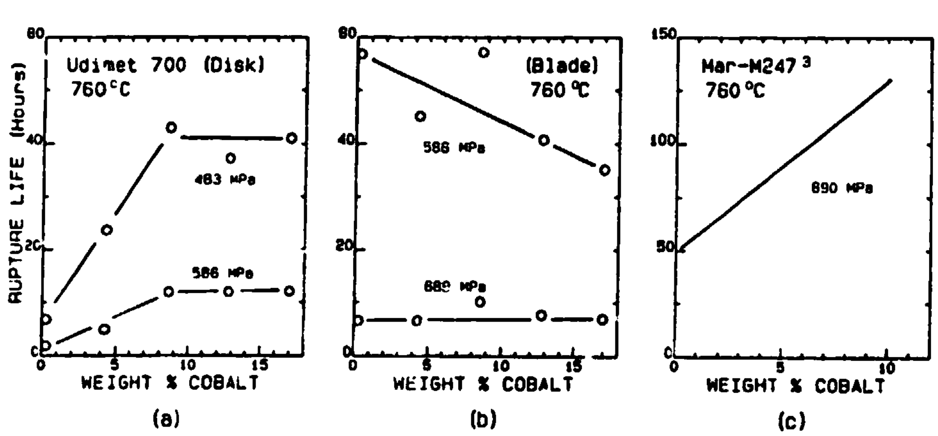


Room temperature tensile properties of Udimet 700[2] [both (a) blade and (b) disk heat treated] and Waspaloy[4] as a function of cobalt content.





Minimum or steady state creep rates of Udimet 700[2] [both (a) disk and (b) blade] and and Mar-M247[3] at 760°C.



Stress Rupture Life of Udimet 700[2] [both (a) disk and (b) blade] and Mar-M247[3] at 760°C.

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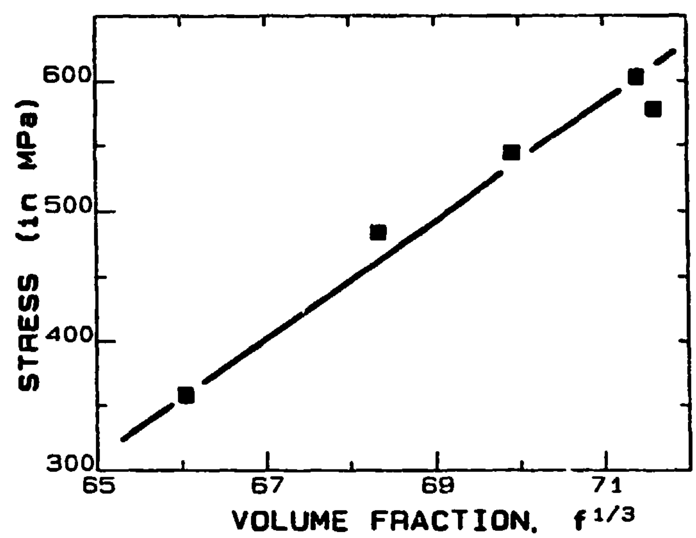
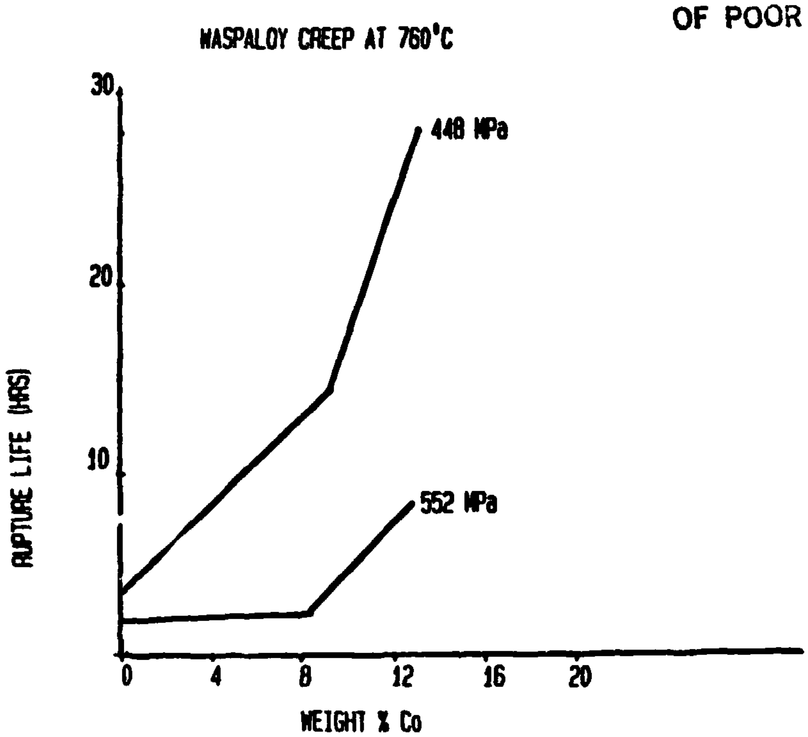


Plot of the Relation Between the Fine γ' Volume Fraction and the Stress Required for a Steady State Creep Rate of 5×10^{-7}cm/cm per sec in the Disk Alloys at 760°C

Creep Resisting Stress Parameters for Various Nickel-Base Superalloys at 760°C (26)

Alloy	σ_p (MPa)	k	σ_s (MPa)	n
TD-Ni	13.1	---	-----	8.0
IN MA754	169.0	0.51	28.6	19.6
Udimet 700	236.2	0.93	226.7	7.6
Nimonic 115	383.0	0.87	123.3	15.0
Mar M 200	465.6	0.87	227.8	12.5
IN MA6000E	466.0	0.72	68.8	24.1



U-700

TENSILE PROPERTIES = F (Co)

DISK CREEP AND STRESS RUPTURE = F (Strengthening γ') = F (Co)

BLADE CREEP AND STRESS RUPTURE = F (TOTAL γ') $\neq$ F (Co)

ACTION: HEAT TREATMENT TO RAISE STRENGTHENING γ' FRACTION IN LOW COBALT U-700

WASPALOY

TENSILE PROPERTIES = F (Co)

CREEP AND STRESS RUPTURE = F (total γ', SFE) = F (Co)

ACTION: Al/Ti VARIATIONS AND SFE STUDIES

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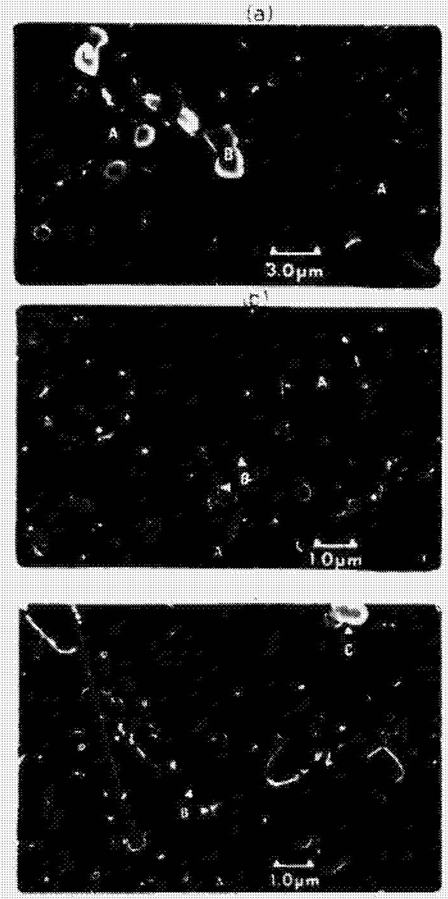
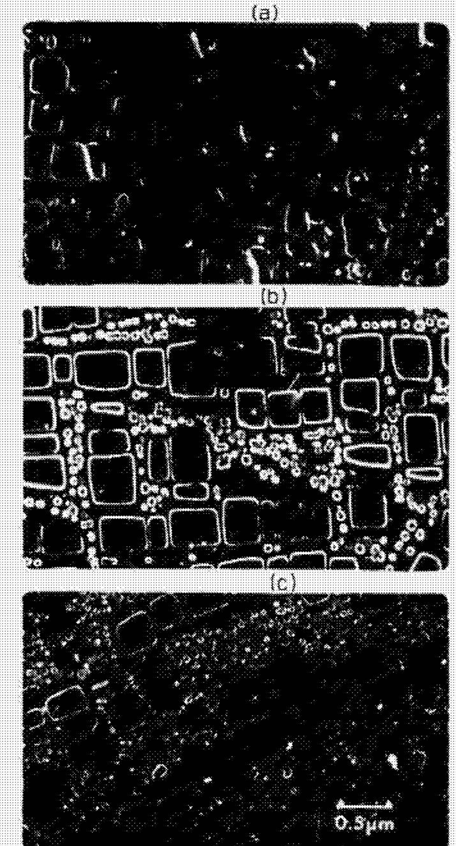
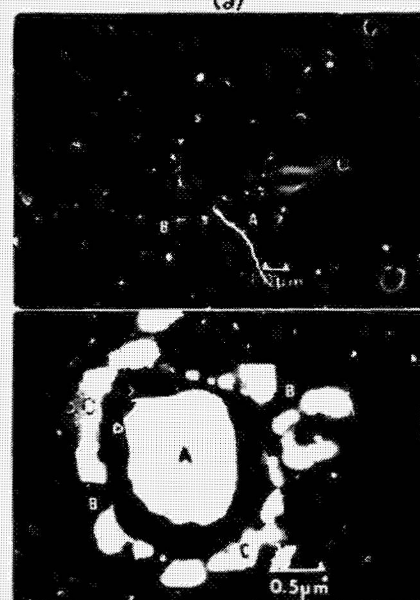


SEM Micrographs of Typical Disk Heat Treated Material with (a) 0. Cobalt, (b) 8.6 Cobalt and (c) 17.0 Cobalt. Markers on the Micrographs Denote (A) Undissolved Gamma Prime, (B) $M_{23}C_6$ Carbides and (C) Primary MC Carbides as determined by EDS.



SEM Micrographs of Fine and Ultrafine γ' in the (a) 0. Cobalt, (b) 8.8 Cobalt and (c) 17.0 Cobalt Blade Heat Treated Material

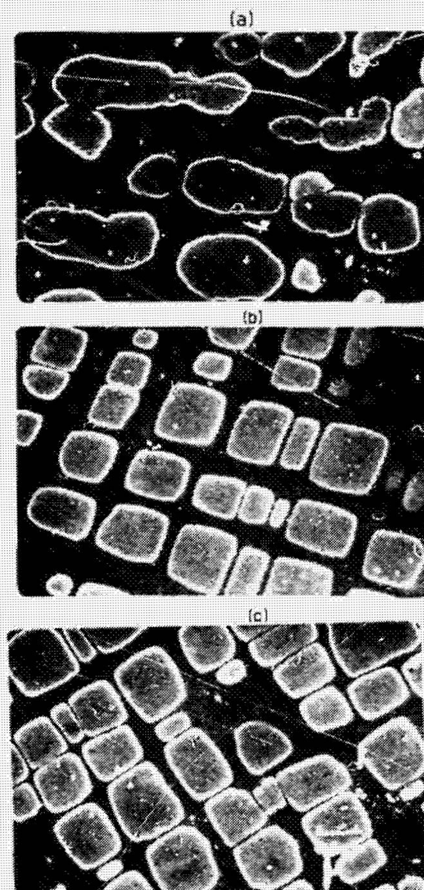
(a)



(b)

SEM Micrographs of Udimet 700 Disk Material Overaged at 815°C for 1000 Hours. (Fig. 12a) Precipitation of Sigma (S) and $M_{23}C_6$ (C) at Grain Boundaries and Around Undissolved γ'. (Fig. 12b) MC Carbide (A) Decomposing into γ' (B) and a Ring of $M_{23}C_6$ (C).

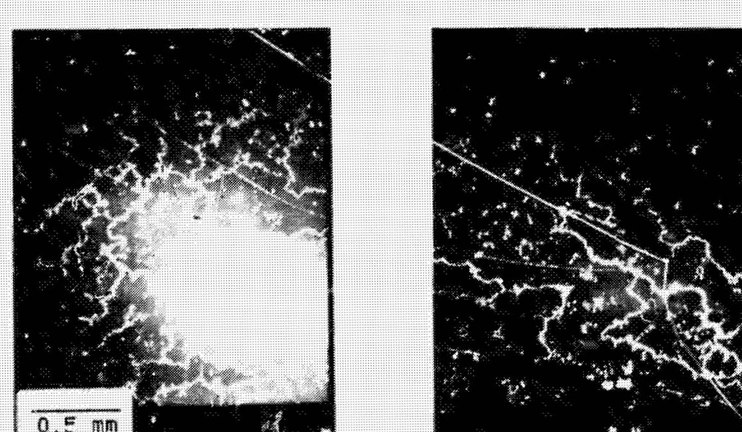
SEM Micrograph of Feathery Sigma Phase (S) in Disk Material after Overaging.



SEM Micrographs of γ' after Coarsening at 982°C for 250 Hours. Note the Elongated γ' in the Cobalt-Free Alloy (a) and the Cubic γ' in (b), 8.8 Cobalt and (c) the 17.0 Cobalt Alloys.

"As Rolled" Microstructure of Low Cobalt Nimonic 115

0.0 % Cobalt 5.2 % Cobalt

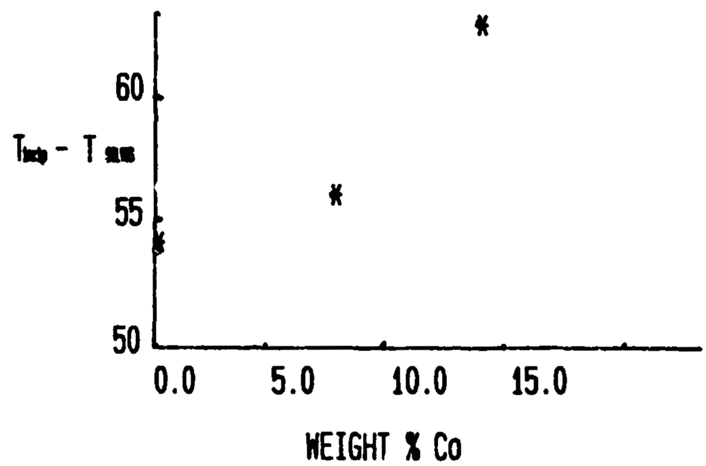


NIMONIC 115

CONVENTIONAL ROLLING TEMPERATURE < M_6C SOLVUS = F (Co)

ACTION: 1.) LOWERING CARBON FROM 0.15% TO 0.07%
2.) INCREASING THE ROLLING TEMPERATURE

U-720 DTA RESULTS



ACTION: ????

1983 PROGRAM

(TESTING OF UNDERSTANDING'S OBTAINED IN 1982)

1.) BASE ON FINDING'S OF CREEP AND STRESS RUPTURE DEPENDENCE ON VOLUME FRACTION γ' FOR U-700
**** IMPROVE CREEP AND STRESS RUPTURE OF LOW COBALT DISK ALLOYS THROUGH HEAT TREATMENT

2.) BASE ON FINDING THAT HIGH MATRIX CONTENT WASPALOY IS BOTH SFE AND γ' FRACTION SENSITIVE
**** IMPROVE CREEP AND STRESS RUPTURE THROUGH Al/Ti VARIATION AND TO DETERMINE SFE IN γ MATRIX WASPALOY

3.) BASED ON FINDING THAT ROLLING TEMPERATURE IS BELOW M_6C SOLVUS FOR LOW COBALT N-115
**** LOWER CARBON CONTENTS
**** ROLL ABOVE M_6C SOLVUS

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EFFECT OF COBALT ON MICROSTRUCTURE AND MICROCHEMISTRY OF NICKEL-BASE SUPERALLOYS

John Radavich and Mayer Engel
Purdue University
School of Materials Engineering
West Lafayette, Indiana 47907

UDIMET 700

As cobalt is removed:

- Total wt. fraction at γ' is relatively unaffected.
- Lattice parameters of the γ' and γ matrix decrease

  a_0 coarse γ' > a_0 fine γ'

- γ/γ' lattice mismatch increases.
- Dominant carbide shift:

  TiC $\longrightarrow$ g.b. $M_{23}C_6$ $\longrightarrow$ massive $M_{23}C_6$

  Massive $M_{23}C_6$ carbides

  - Found only in low Co alloys
  - Often occurring in clusters or stringers

- The γ, γ', $M_{23}C_6$ and M_3B_2 phases contain lesser amounts of Co.

Effects of long time aging (LTA)

- Sigma phase forms in 8.6, 12.8 and 17.0 wt% Co alloys, becoming less abundant as Co is removed.
- Additional precipitation of g.b. $M_{23}C_6$ carbides in all alloys aside from Co-free version..

Microchemistry (SEM/EDAX analysis) of MC, $M_{23}C_6$ and M_3B_2 phases

- No Co in MC carbides
- $[Co]_{M_{23}C_6} > [Co]_{M_3B_2}$
- Analysis of Mo, Ti, Cr and Ni also investigated as a function of alloy Co content.

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NIMONIC - 115

As Cast

- MC - dominant phase
- $M_{23}C_6$ - trace in low Co alloys.

As Rolled

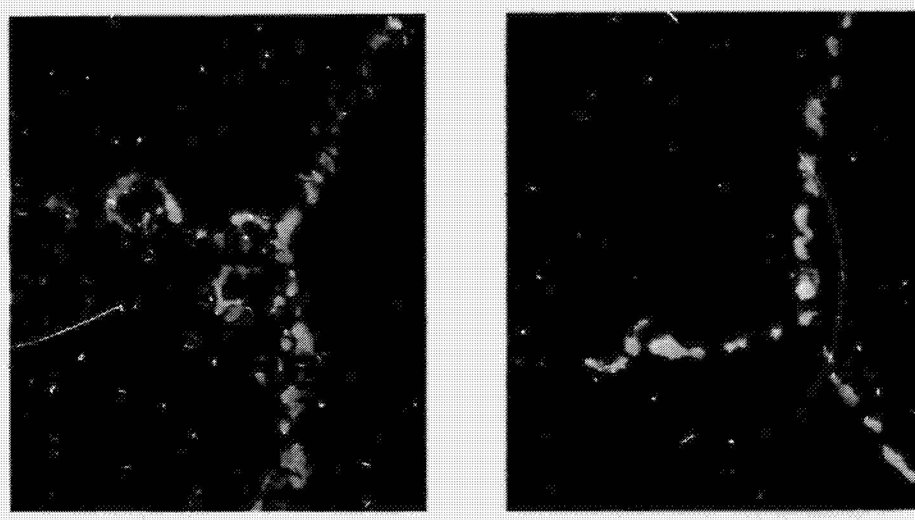
- Low Co alloys (0 + 5 wt.% Co) did not roll due to continuous g.b. $M_{23}C_6$ carbides
- High Co alloys (10 + 14 wt.% Co) rolled as $M_{23}C_6$ uniformly dispersed throughout the material.

BLADE H.T.

- MC - dominant phase in high Co alloys.

SEM Observations

- Low Co decreases matrix solubility for carbon and boron
- Low Co affects the as-cast microstructure
- Low Co affects the size and amount of γ' produced during casting and/or heat treatment.



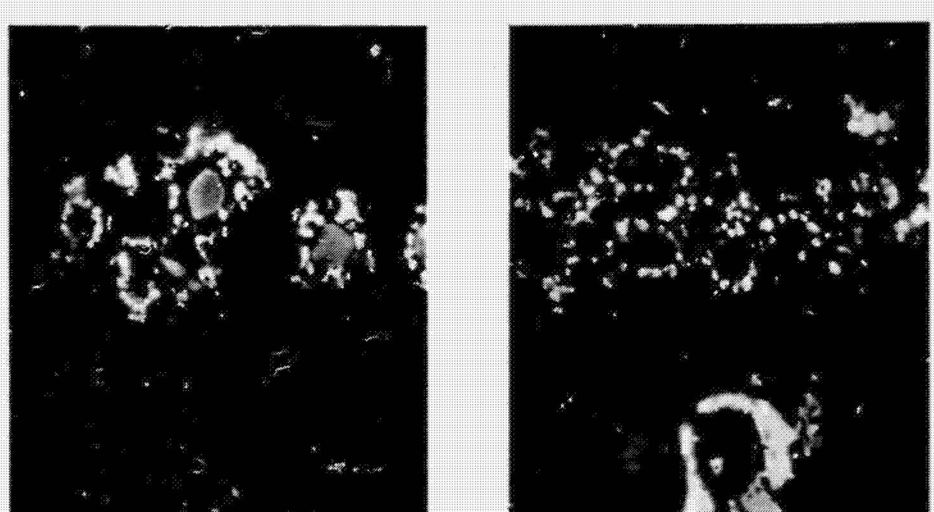
0% Co 17% Co

U-700 Grain Boundaries

D-6 HT + 1500°F/50 hr

5000X

Fig. 1



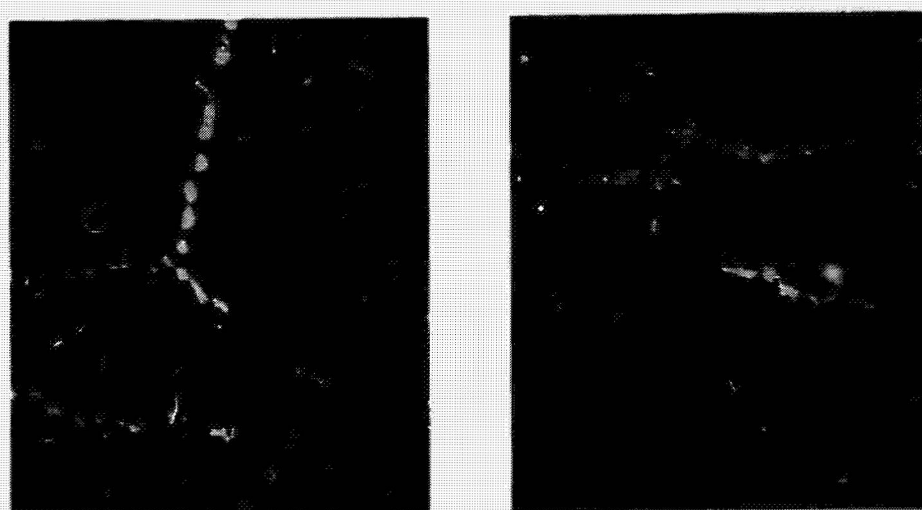
Boride MC

U-700 0% Co

D-6 HT + 1550°F/50 hr

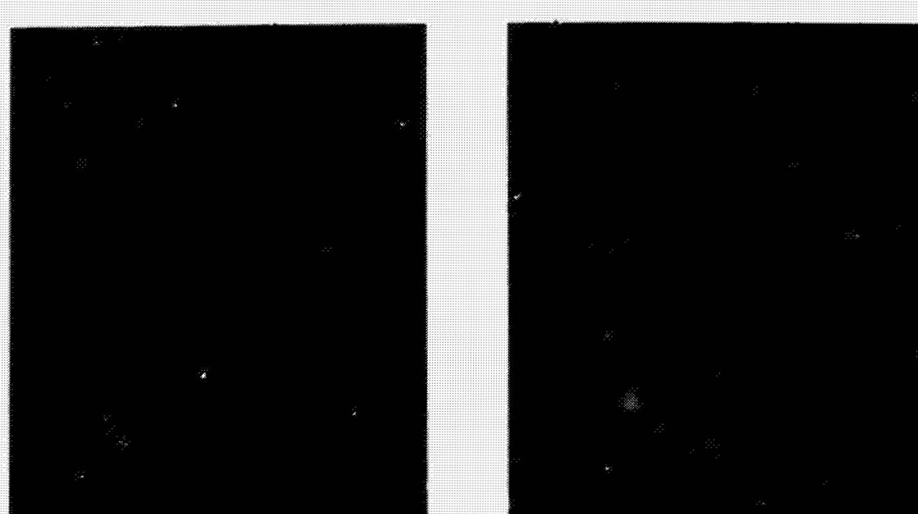
5000 X

Fig. 2



U-700 17% Co
D-S HT + 1550°F/50 hr
5000 X

Fig. 3



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NIM 115 AS CAST 0% Co

Fig. 4

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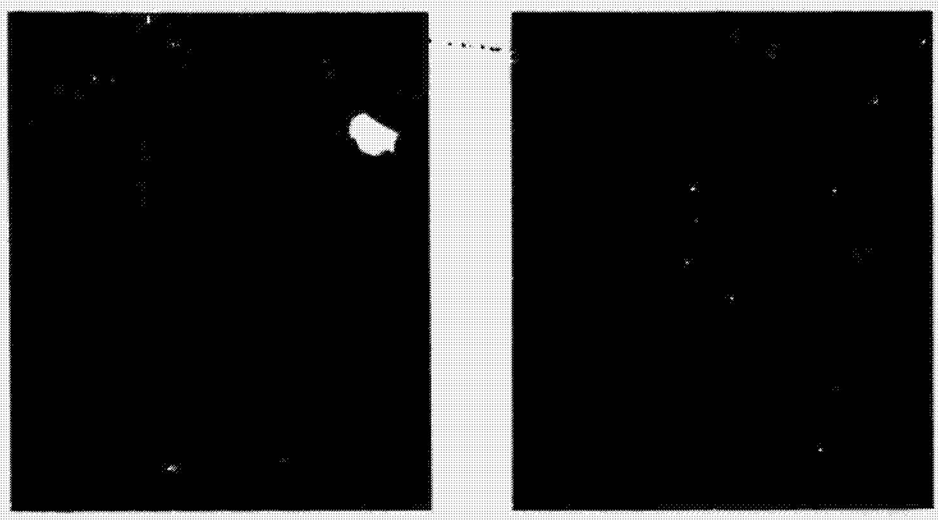
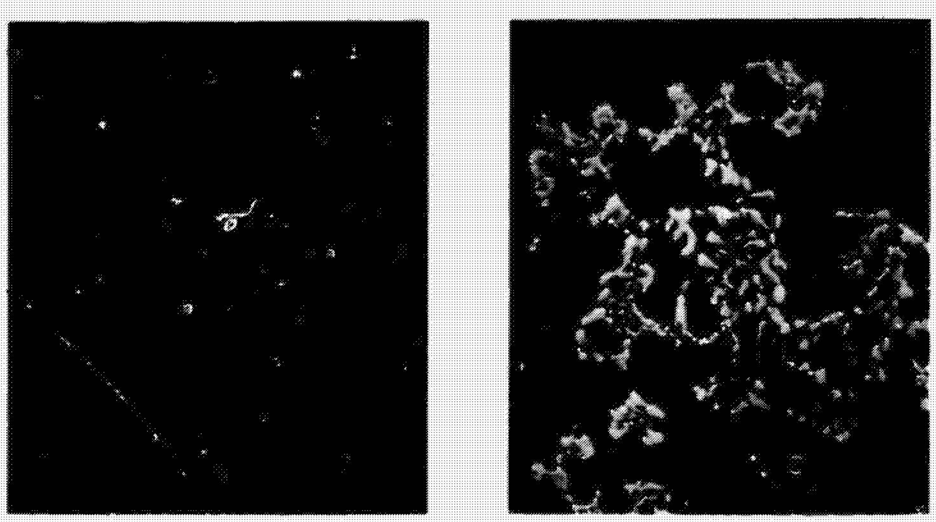


1000X 3000X

NIM 115 AS CAST 10% Co

Fig. 5



300X 3000X

NIM 115 AS ROLLED 0% Co

Fig. 6

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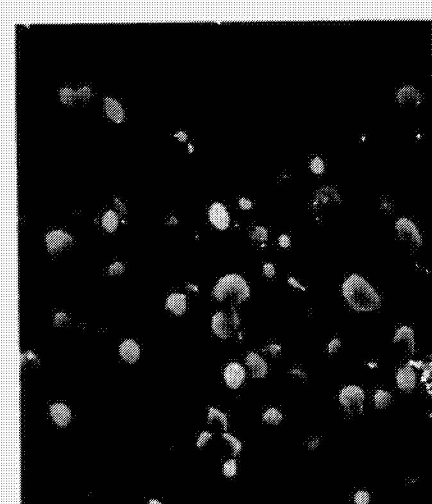
3000X

NiM 115 AS ROLLED 5% Co

Fig. 7



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NIM 115 AS ROLLED 10% Co

Fig. 8

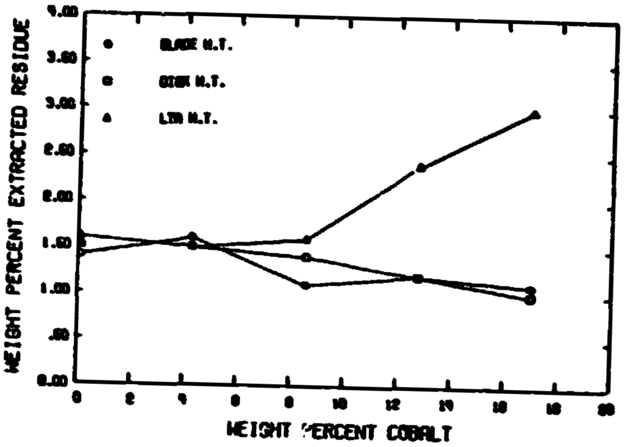

FIGURE XX. WEIGHT PERCENT EX-1(HCL) RESIDUE IN U-700 BLADE, DISK AND LTA MATERIAL

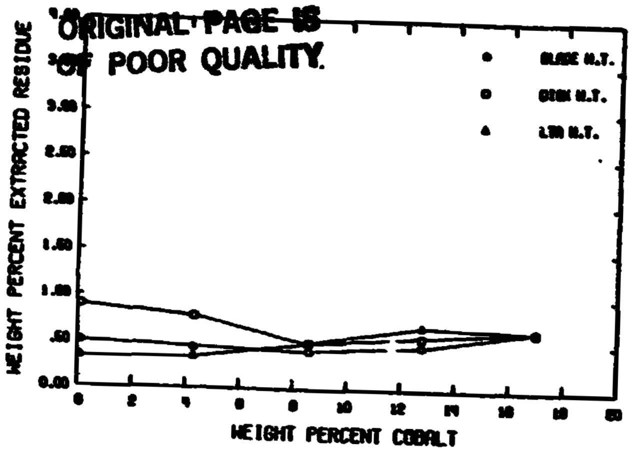

FIGURE XX. WEIGHT PERCENT EX-2(BR) RESIDUE IN U-700 BLADE, DISK AND LTA MATERIAL

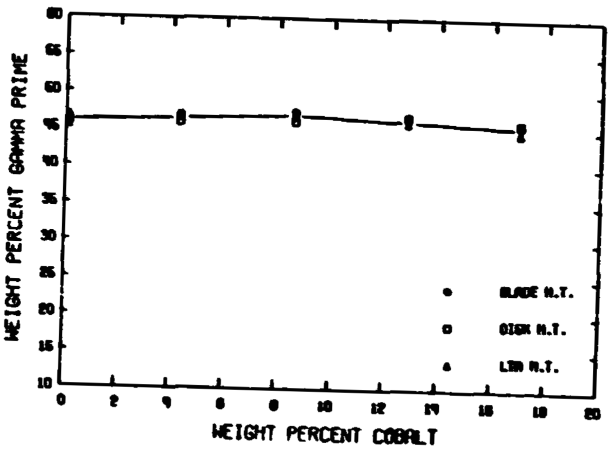

FIGURE XX. WEIGHT PERCENT GAMMA PRIME IN U-700 BLADE, DISK AND LTA MATERIAL

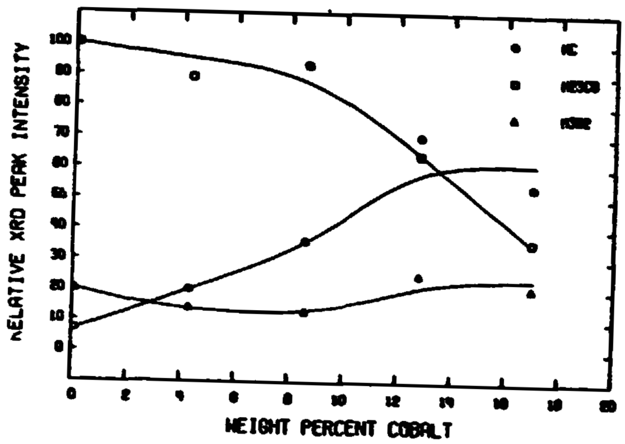

FIGURE XX. RELATIVE XRD PEAK INTENSITIES OF U-700 BLADE EX-1(HCL) EXTRACTED RESIDUES

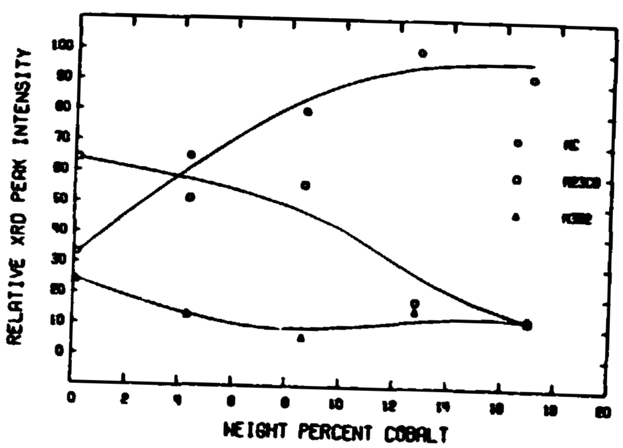

FIGURE XX. RELATIVE XRD PEAK INTENSITIES OF U-700 BLADE EX-2(BR) EXTRACTED RESIDUES

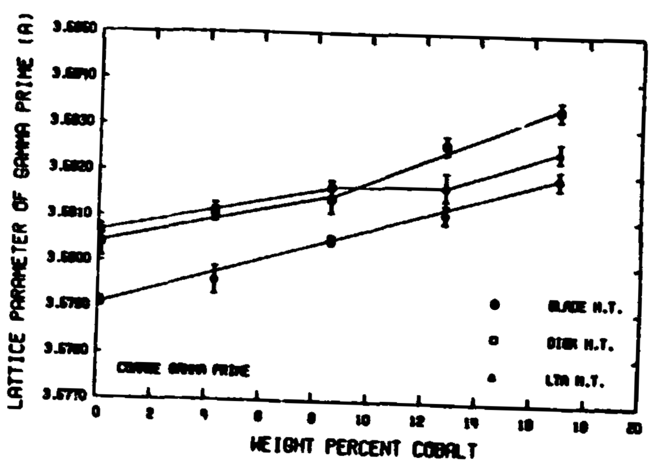

FIGURE XX. LATTICE PARAMETERS OF COARSE GAMMA PRIME IN U-700 BLADE, DISK AND LTA MATERIAL

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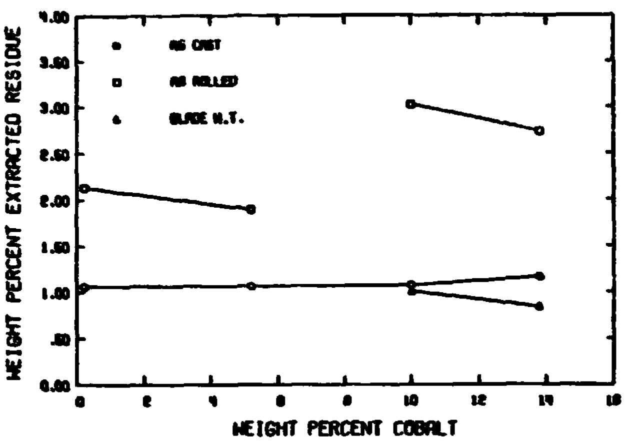

FIGURE XX. WEIGHT PERCENT EX-1(HCL) RESIDUE IN N-115 AS CAST, AS ROLLED AND BLADE MATERIAL

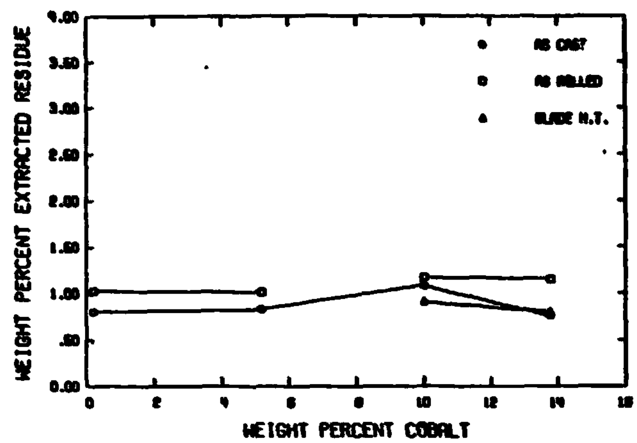

FIGURE XX. WEIGHT PERCENT EX-2(BR) RESIDUE IN N-115 AS CAST, AS ROLLED AND BLADE MATERIAL

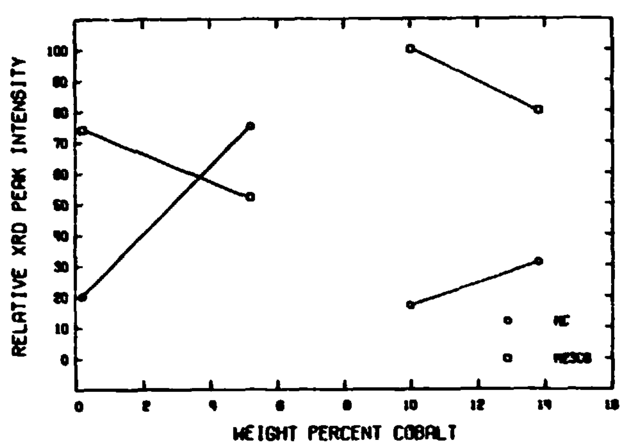

FIGURE XX. RELATIVE XRD PEAK INTENSITIES OF N-115 AS ROLLED EX-1(HCL) EXTRACTED RESIDUES

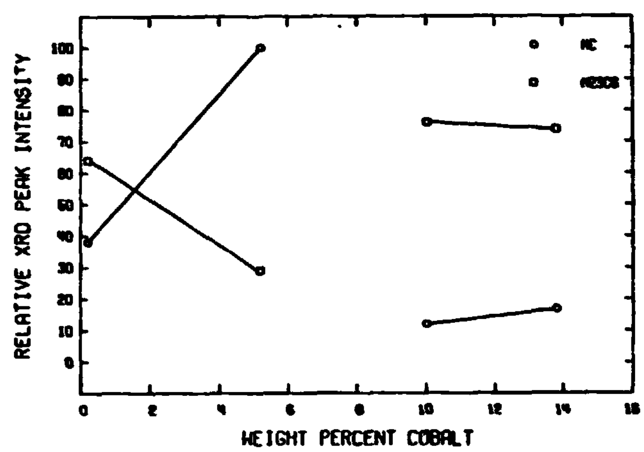

FIGURE XX. RELATIVE XRD PEAK INTENSITIES OF N-115 AS ROLLED EX-2(BR) EXTRACTED RESIDUES

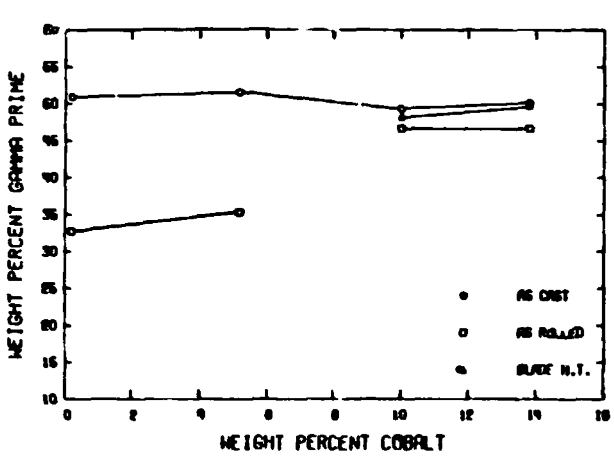

FIGURE XX. WEIGHT PERCENT GAMMA PRIME IN N-115 AS CAST, AS ROLLED AND BLADE MATERIAL

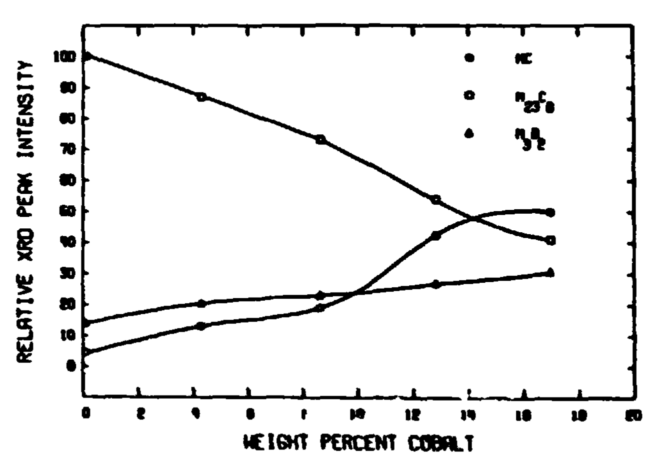

FIGURE 18. RELATIVE XRD PEAK INTENSITIES OF U-700 DISK EX-1(HCL) EXTRACTED RESIDUES

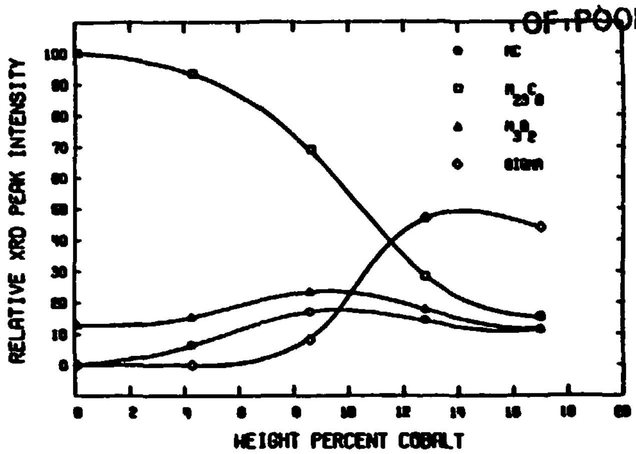

FIGURE 20. RELATIVE XRD PEAK INTENSITIES OF U-700 LTA EX-1(HCL) EXTRACTED RESIDUES

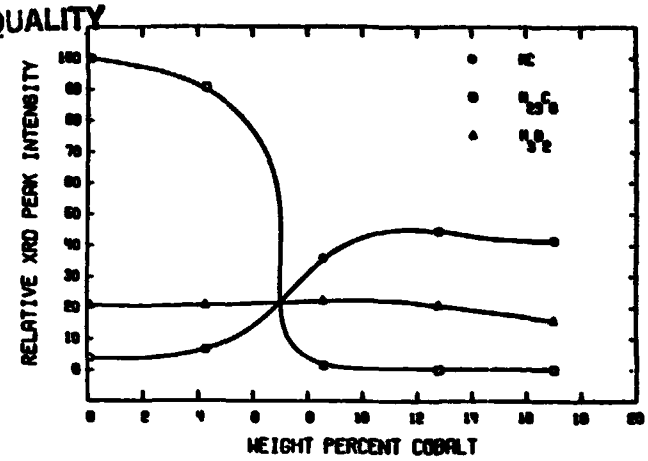

FIGURE 21. RELATIVE XRD PEAK INTENSITIES OF U-700 DISK (EX-2(BR)) EXTRACTED RESIDUES

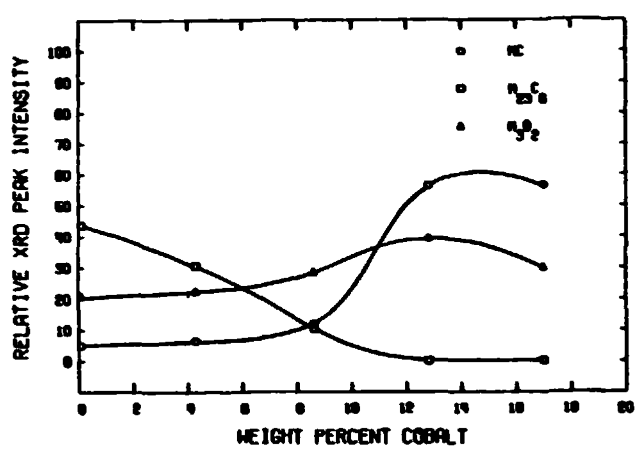

FIGURE 22. RELATIVE XRD PEAK INTENSITIES OF U-700 LTA (EX-2(BR)) EXTRACTED RESIDUES

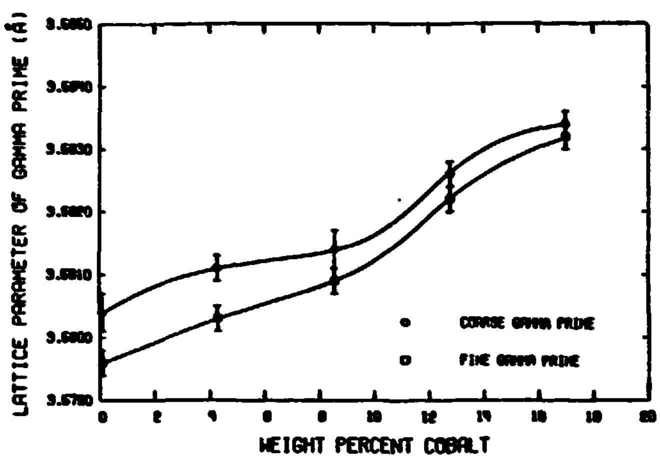

FIGURE 23. LATTICE PARAMETERS OF GAMMA PRIME IN U-700 DISK MATERIAL

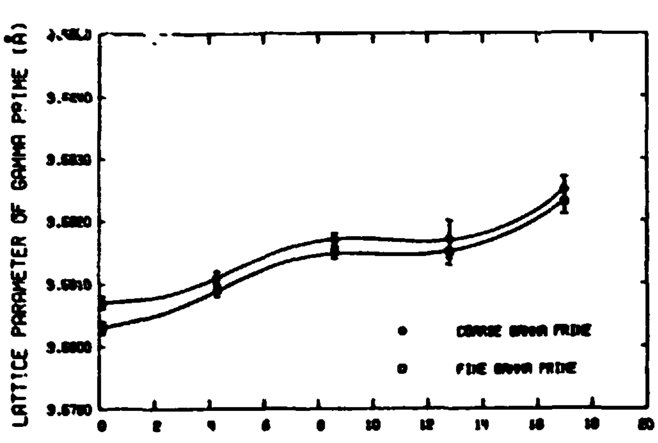

FIGURE 24. LATTICE PARAMETERS OF GAMMA PRIME IN U-700 LTA MATERIAL

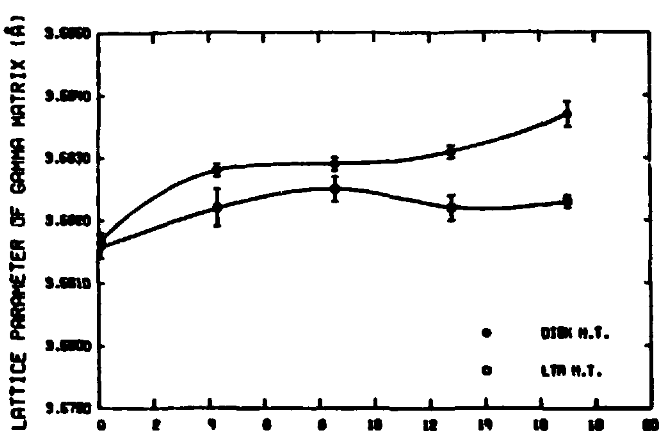

FIGURE 25. LATTICE PARAMETERS OF GAMMA MATRIX IN U-700 DISK AND LTA MATERIAL

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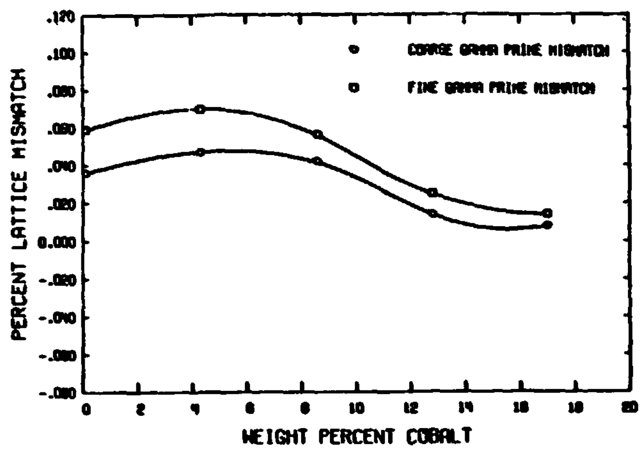

FIGURE 26. PERCENT LATTICE MISMATCH OF GAMMA/GAMMA PRIME IN U-700 DISK MATERIAL

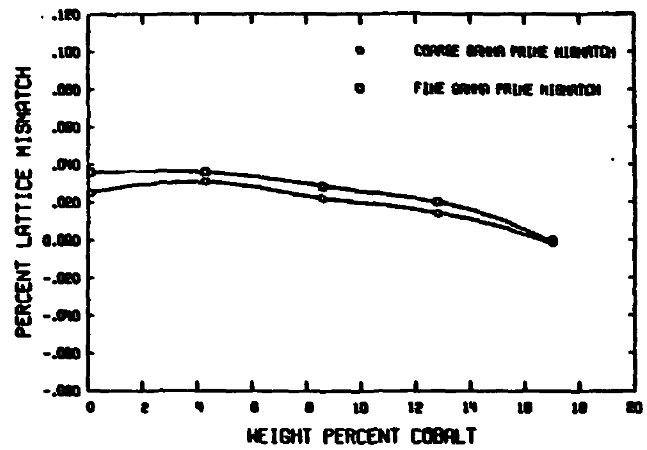

FIGURE 27. PERCENT LATTICE MISMATCH OF GAMMA/GAMMA PRIME IN U-700 LTA MATERIAL

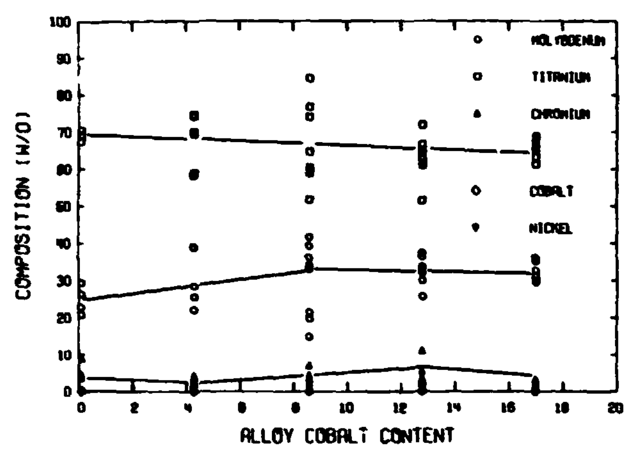

FIGURE 1C. SEM/EDAX ANALYSIS OF MC CARBIDE IN U-700 DISK MATERIAL (STD ANAL. AUTO BKG)

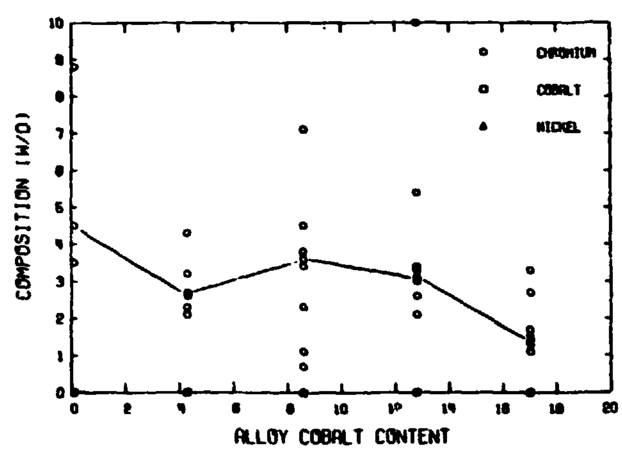

FIGURE 1C. SEM/EDAX ANALYSIS OF MC CARBIDE IN U-700 DISK MATERIAL (STD ANAL. AUTO BKG)

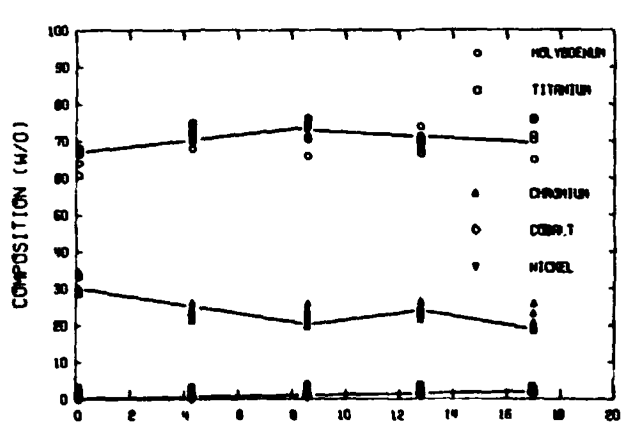

FIGURE 28. SEM/EDAX ANALYSIS OF M3B2 BORIDE IN U-700 DISK MATERIAL (STD ANAL. AUTO BKG)

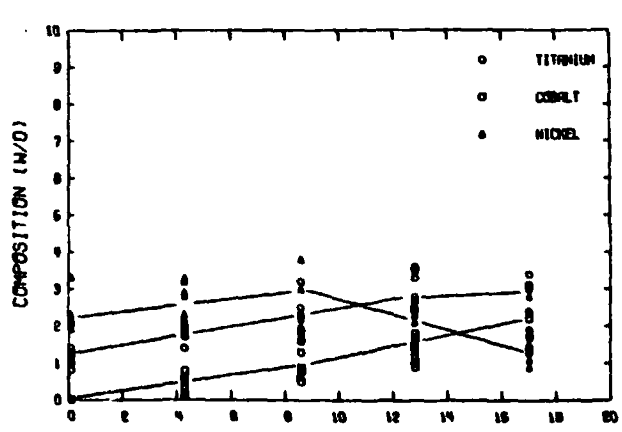

FIGURE 28. SEM/EDAX ANALYSIS OF M3B2 BORIDE IN U-700 DISK MATERIAL (STD ANAL. AUTO BKG)

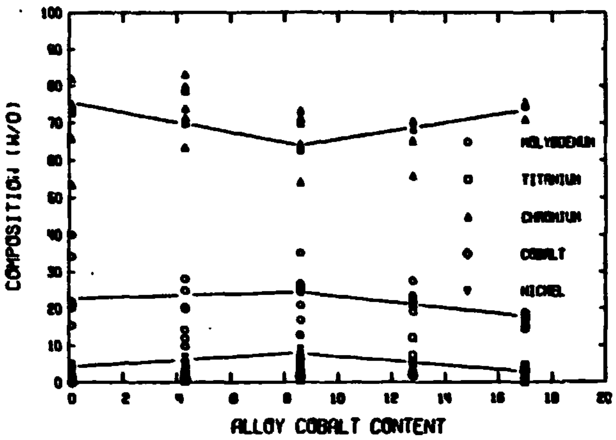

FIGURE 38. SEM/EDAX ANALYSIS OF M23C6 CARBIDE IN U-700 DISK MATERIAL (STD ANAL. AUTO BKG)

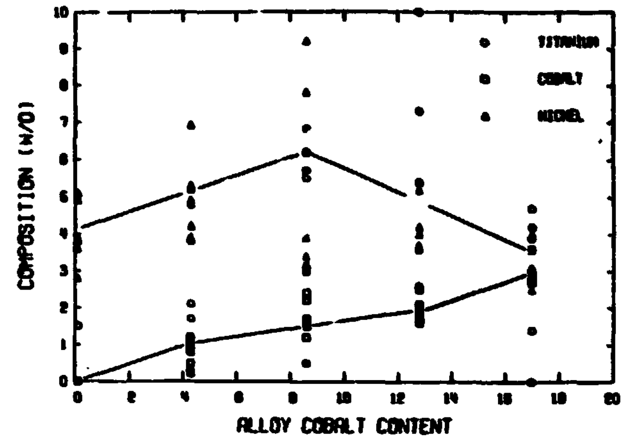

FIGURE 39. SEM/EDAX ANALYSIS OF M23C6 CARBIDE IN U-700 DISK MATERIAL (STD ANAL. AUTO BKG)

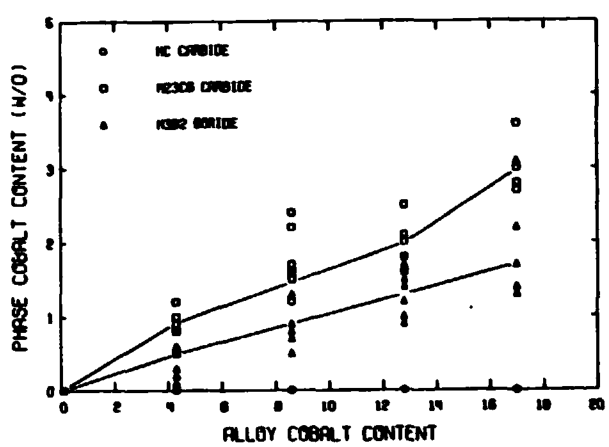

FIGURE 40. SEM/EDAX ANALYSIS OF COBALT IN MC, M23C6 AND M3B2 IN U-700 DISK MATERIAL(STD)

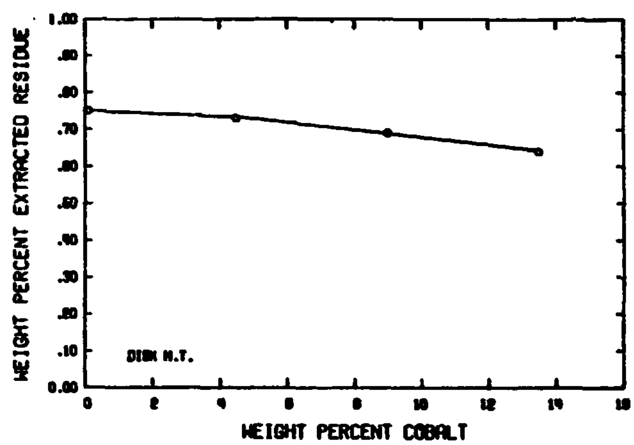

FIGURE XX. WEIGHT PERCENT EX-1(HCL) RESIDUE IN WASPALOY DISK MATERIAL

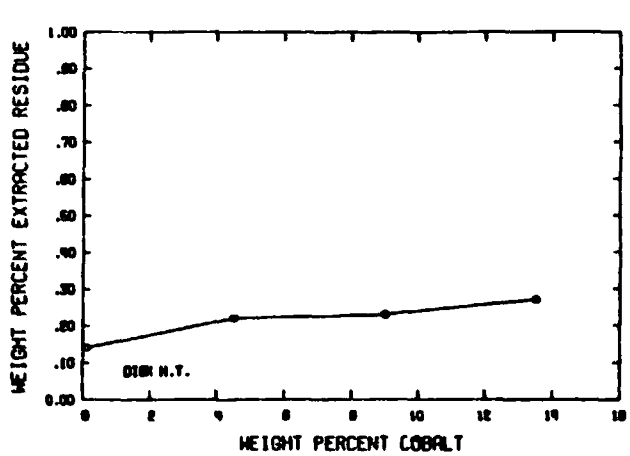

FIGURE XX. WEIGHT PERCENT EX-2(BR) RESIDUE IN WASPALOY DISK MATERIAL

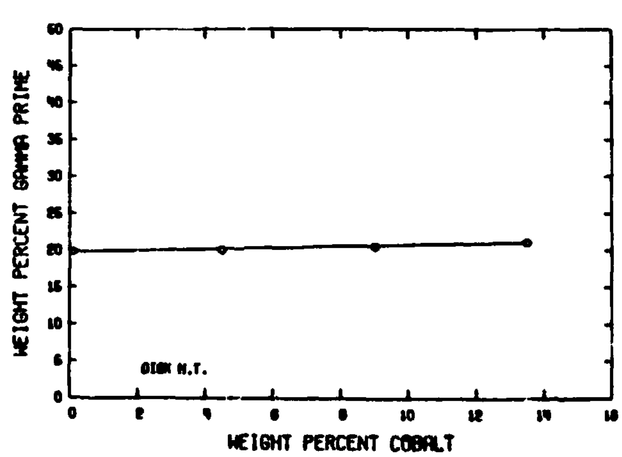

FIGURE XX. WEIGHT PERCENT GAMMA PRIME IN WASPALOY DISK MATERIAL

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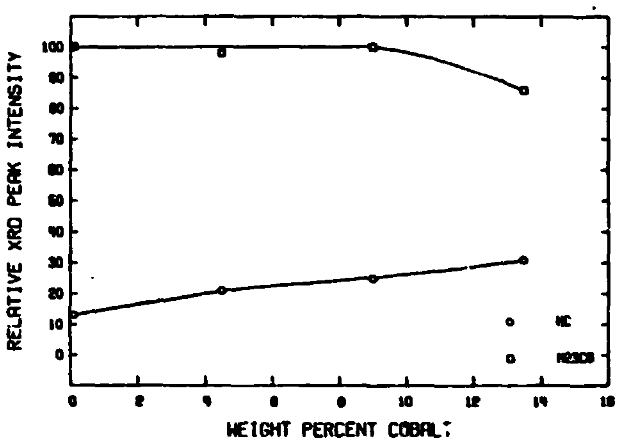

FIGURE XX. RELATIVE XRD PEAK INTENSITIES OF WASPALOY DISK EX-1(HCL) EXTRACTED RESIDUES

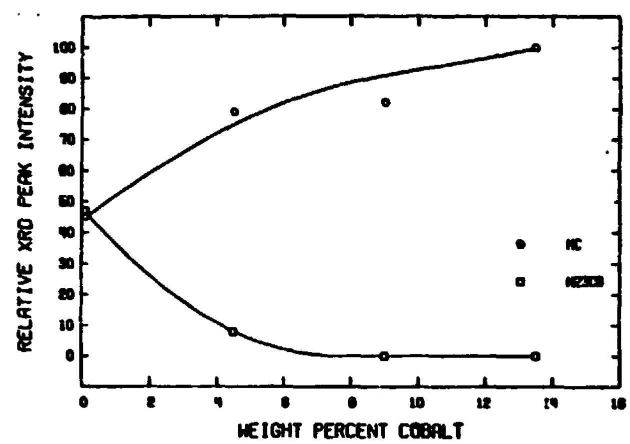

FIGURE XX. RELATIVE XRD PEAK INTENSITIES OF WASPALOY DISK EX-2(BR) EXTRACTED RESIDUES

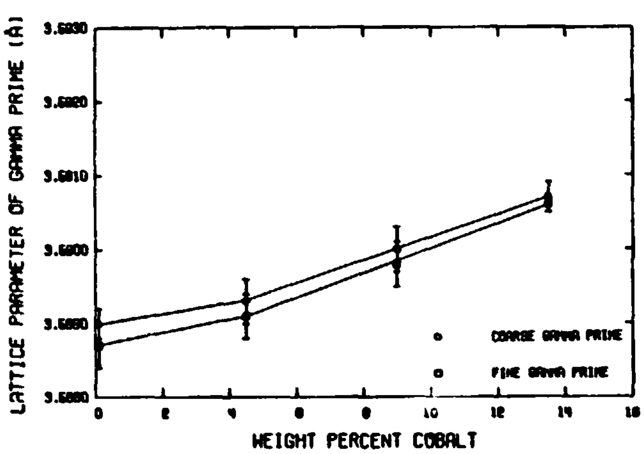

FIGURE XX. LATTICE PARAMETERS OF GAMMA PRIME IN WASPALOY DISK MATERIAL

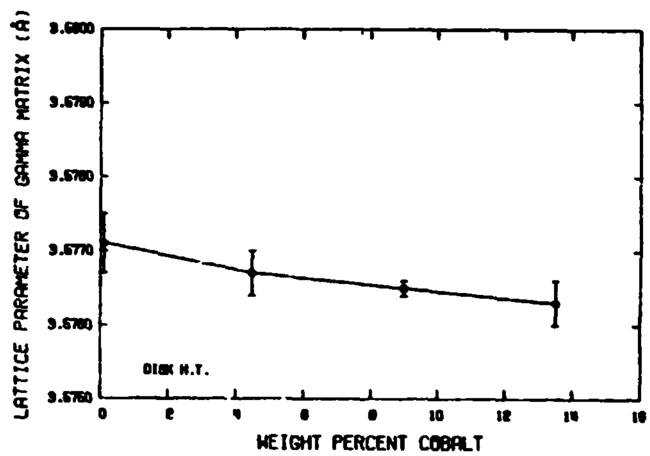

FIGURE XX. LATTICE PARAMETERS OF GAMMA MATRIX IN WASPALOY DISK MATERIAL

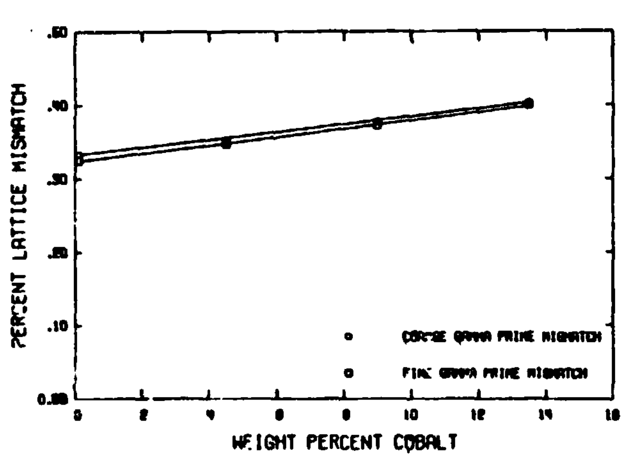

FIGURE XX. PERCENT LATTICE MISMATCH OF GAMMA/GAMMA PRIME IN WASPALOY DISK MATERIAL

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EFFECT OF REDUCED COBALT CONTENTS ON HOT ISOSTATICALLY PRESSED POWDER METALLURGY U-700 ALLOYS

Fredric H. Harf
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

Prealloyed powders of Udimet 700 (U-700) alloys in which the cobalt content was reduced from the normal 17 to 19 percent to 12.7, 8.6, 4.3, and 0 percent were hot isostatically pressed (HIP) into billets. These billets were given heat treatments appropriate for turbine disks, namely partial solutioning at temperatures below the gamma prime solvus and four-step aging treatments. Chemical analyses, metallographic examinations, and X-ray diffraction measurements have been performed on these materials.

Reducing cobalt in powder metallurgy (P-M) U-700 had only minor effects on gamma prime content and on room temperature and 650° C tensile properties. Creep-rupture lives at 650° C reached a maximum at the 8.4 percent cobalt concentration while at 760° C a maximum in life was reached at the 4.3-percent cobalt level. Minimum creep rates increased with decreasing cobalt content in P-M U-700 at both test temperatures.

Extended exposures at 760° and 815° C resulted in decreased tensile strengths and rupture lives for all alloys. Evidence of sigma phase formation was also found. The effects of minor adjustments in chromium, molybdenum, aluminum, and titanium to substitute for reduced cobalt levels in P-M U-700 will be determined in a continuation of this program.

COBALT IN POWDER METALLURGY U-700

- DETERMINE EFFECTIVENESS OF COBALT ON
  - PROPERTIES
  - STRUCTURE
- IDENTIFY SUBSTITUTES OTHER THAN NICKEL

COMPOSITION OF HIP P-M U-700 ALLOYS

Co	Cr	Mo	Ti	Al	C	B	Ni
0	15.0	5.00	3.51	4.00	.065	.019	bal
4.3	14.9	4.85	3.53	4.04	07	.020	bal
8.55	14.8	5.00	3.54	4.08	.06	.022	bal
12.7	14.8	5.10	3.57	4.04	.06	.023	bal
17.0	14.8	5.10	3.56	4.04	.06	.026	bal

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HEAT TREATMENT OF HIP P-M U-700 ALLOYS

COBALT CONTENT, %	GAMMA PRIME SOLVUS, °C	PARTIAL SOLUTIONING 4 hr AT °C, OIL QUENCH*
0	1188	1146 (1129)
4.3	1180	1138 (1129)
8.55	1170	1129 (1129)
12.7	1160	1118 (1118)
17.0	1150	1104 (1104)

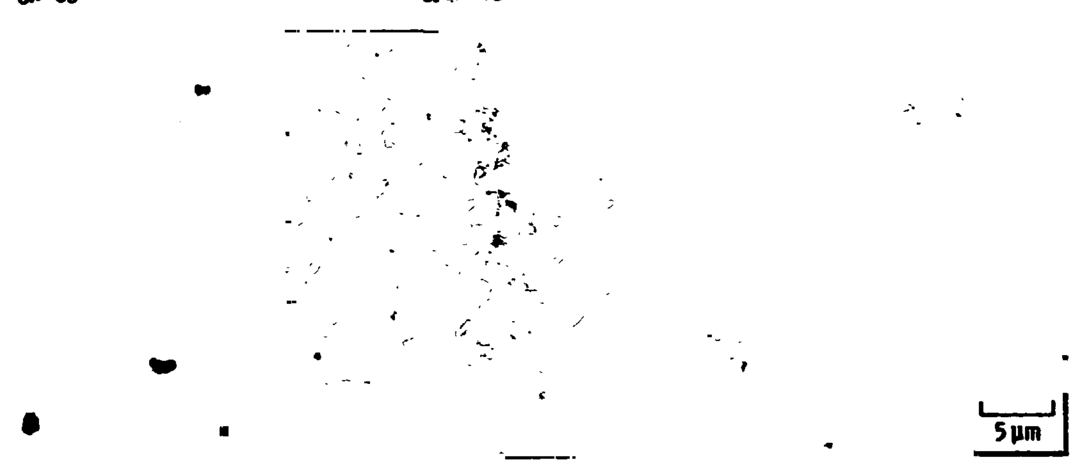
AGING TREATMENT SEQUENCE FOR ALL ALLOYS

870° C-8 hr-AIR COOL
980° C-4 hr-AIR COOL
650° C-24 hr-AIR COOL
760° C-8 hr-AIR COOL

* CAST AND WROUGHT ALLOYS OF THE SAME COMPOSITIONS WERE PARTIALLY SOLUTIONED AT THE TEMPERATURES SHOWN IN PARENTHESES (JARRETT & TIEN, MET. TRANS., 13A, PP. 1021-1032)

MICROSTRUCTURES OF HEAT-TREATED P-M U-700 ALLOYS

0% Co 8.6% Co 17% Co



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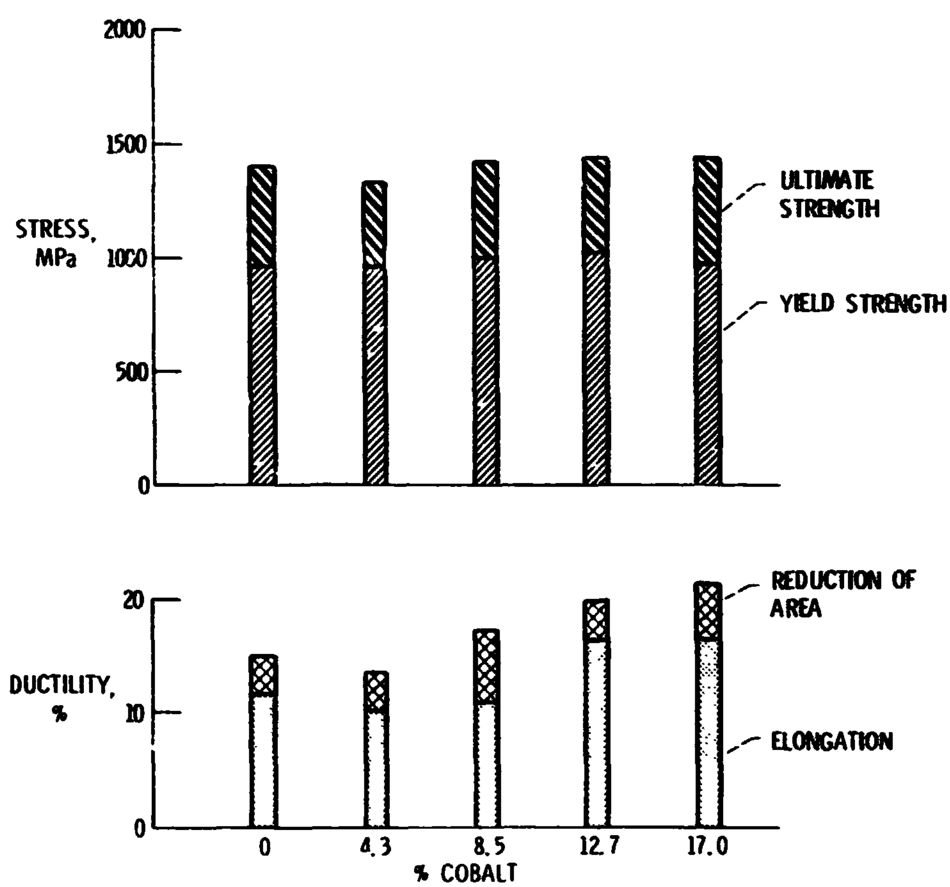
MICROSTRUCTURAL FEATURES OF HEAT-TREATED P-M U-700 ALLOYS

COBALT, %	GAMMA PRIME, wt%	LATTICE PARAMETERS, Å		PERCENT
		GAMMA	GAMMA PRIME	MISMATCH
0	46.7	3.5859	3.5821	0.106
4.3	46.4	3.5860	3.5824	0.100
8.5	46.8	3.5858	3.5835	0.064
12.7	45.8	3.5857	3.5841	0.045
17.0	45.6	3.5851	3.5841	0.006

ROOM TEMPERATURE TENSILE PROPERTIES OF P-M U-700 ALLOYS



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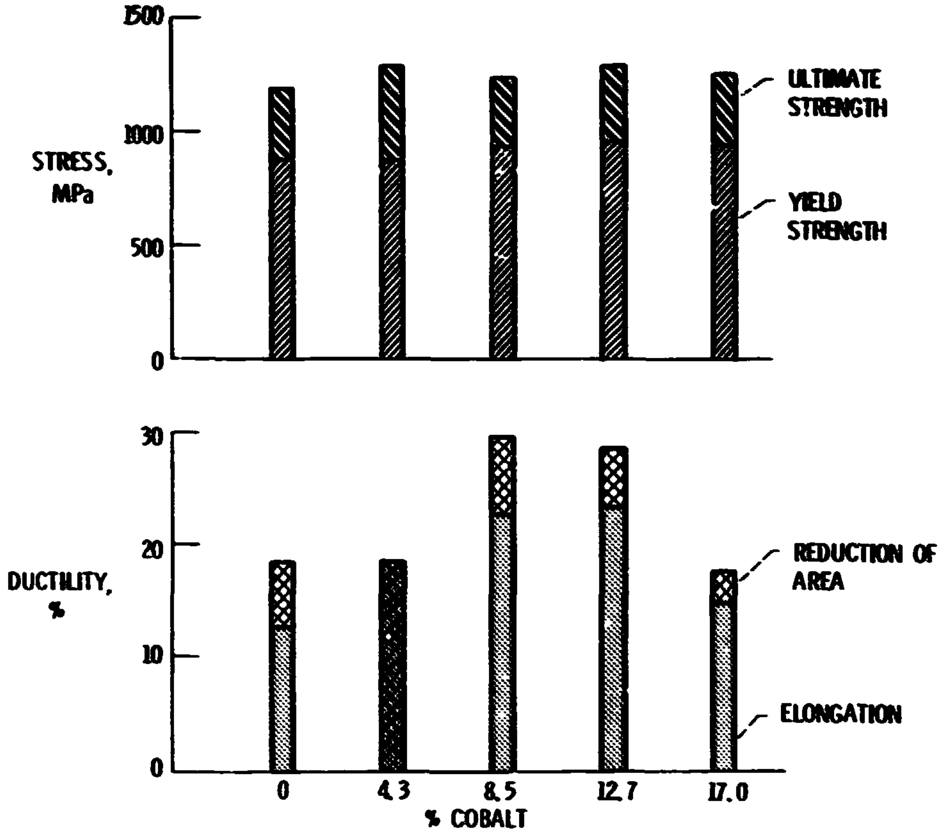
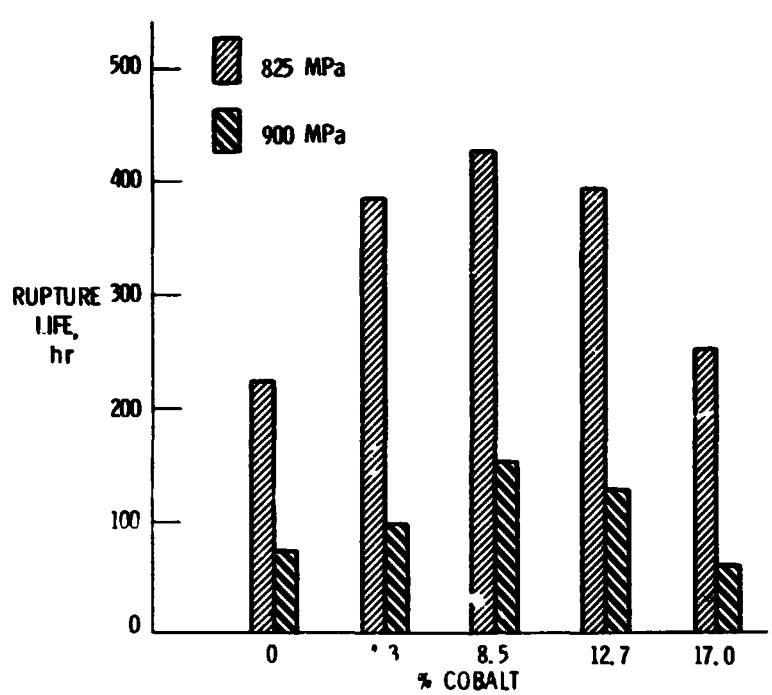
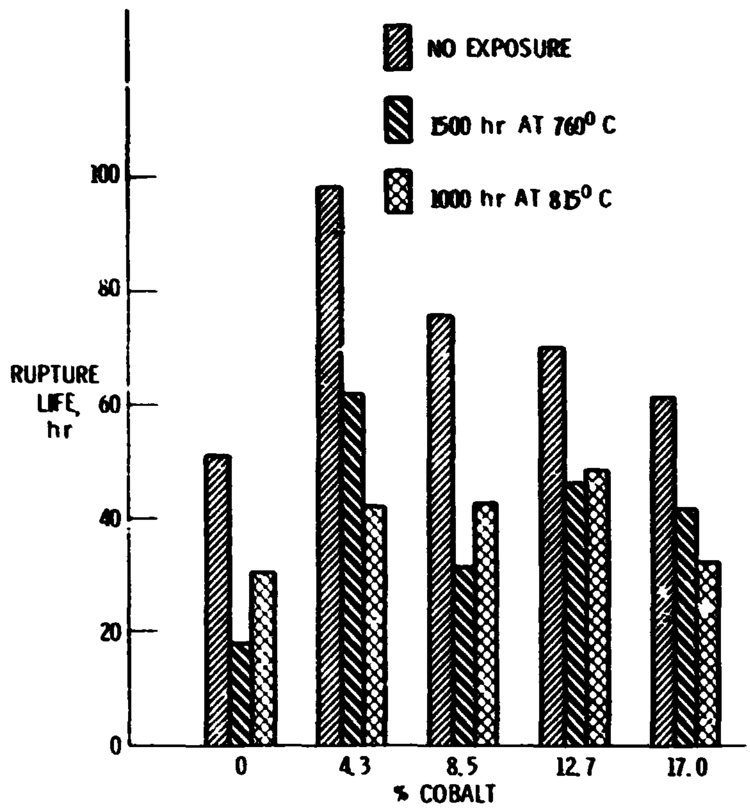
650° C TENSILE PROPERTIES OF P-M U-700 ALLOYS



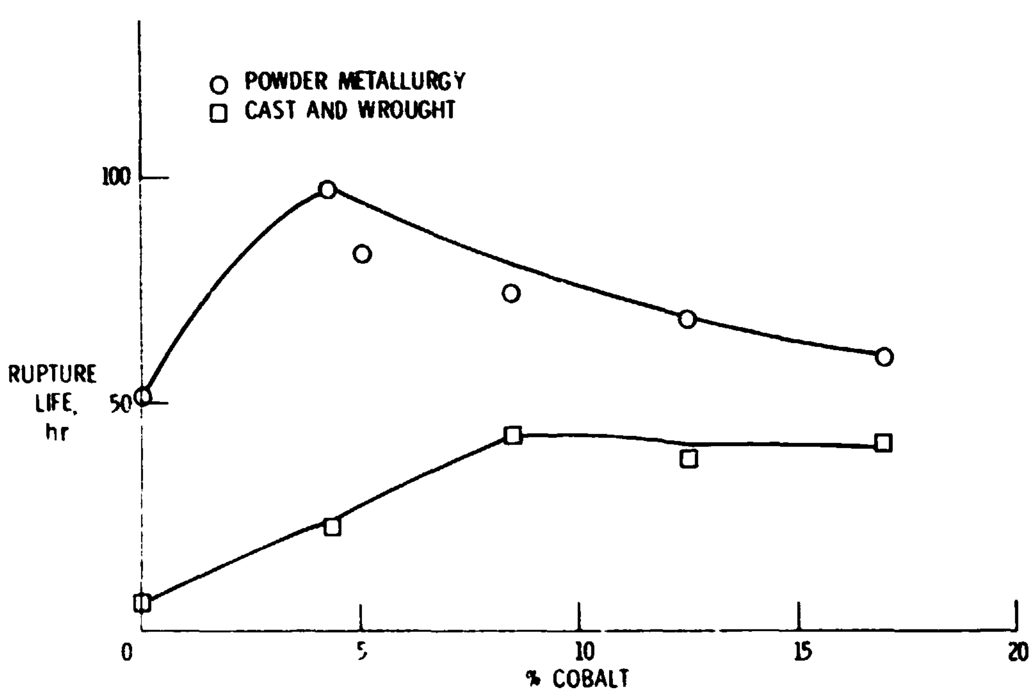
650° C RUPTURE LIVES OF P-M U-700 ALLOYS



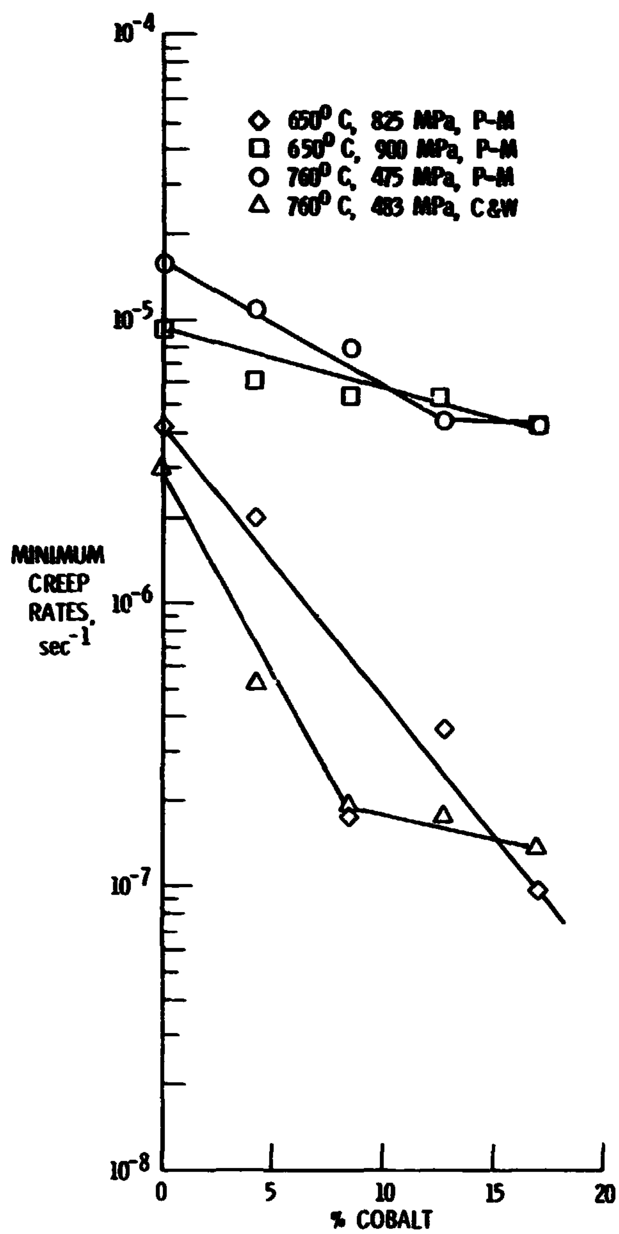
EFFECT OF ELEVATED TEMPERATURE EXPOSURE ON RUPTURE LIVES AT 760° C, 475 MPa



760° C RUPTURE LIVES OF POWDER METALLURGY AND OF CAST & WROUGHT ALLOYS



CREEP BEHAVIOR OF P-M AND CAST-WROUGHT U-700 ALLOYS



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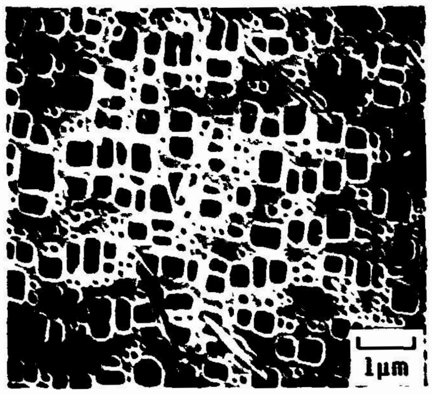
INDICATION OF SIGMA PHASE AFTER LONG TIME EXPOSURE TO ELEVATED TEMPERATURES



17% COBALT ALLOY
1500 hr EXPOSURE AT 845 ^{0} C

POSSIBLE X-RAY INDICATIONS ALSO FOUND IN OTHER ALLOYS AFTER > 500 hr AT 760^{0} C AND 845^{0} C

CS-82-2207

CRITERIA FOR SELECTION OF NEW P-M COMPOSITIONS

1. PREDICTED GAMMA—GAMMA PRIME MISMATCH IN RANGE OF BETTER ALLOYS TESTED
2. PREDICTED GAMMA PRIME CONTENT 45 TO 50%
3. PREDICTED FREE OF SIGMA PHASE

CONCLUSIONS

IN P-M U-700 ALLOYS WITH A DISK HEAT TREATMENT THE PRESENCE OF COBALT

- DOES NOT CHANGE THE AMOUNT OF GAMMA PRIME
- AT LEVELS BETWEEN 4 AND 12 PERCENT PROVIDES BETTER RUPTURE LIVES THAN AT 0% AND AT THE STANDARD 17 TO 19%
- TENDS TO IMPROVE THE MINIMUM CREEP RATE
- DOES NOT IMPROVE TENSILE STRENGTH AT 25^{0} AND 650^{0} C
- HAS A MINOR EFFECT ON TENSILE DUCTILITY AT 25^{0} AND 650^{0} C

N83 11288 D6-26

LOW-COBALT SINGLE CRYSTAL RENÉ 150

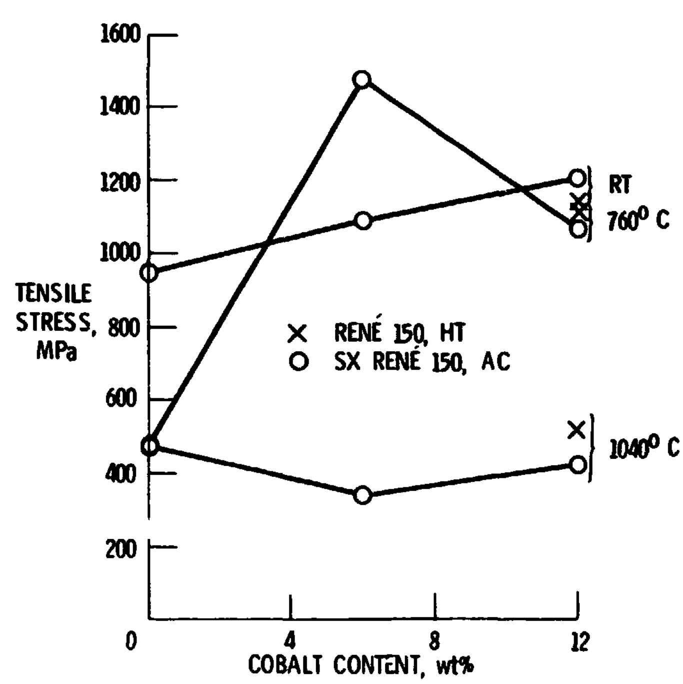
Coulson M. Scheuermann
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

An experimental program is underway at NASA Lewis Research Center to investigate the effects of cobalt content on a single crystal version of the advanced, high gamma prime content turbine airfoil alloy René 150. Cobalt contents under investigation include 12 wt.% (composition level of René 150), 6 wt.%, and 0 wt.%. Preliminary test results are presented and compared with the properties of standard DS René 150. DTA results indicate that the liquidus goes through a maximum of about 1435° C near 6 wt.% Co. The solidus remains essentially constant at 1390° C with decreasing Co content. The gamma prime solvus appears to go through a minimum of about 1235° C near 6 wt.% Co content. Preliminary as-cast tensile and stress-rupture results are presented along with heat treat schedules and future test plans.

ALLOY COMPOSITIONS

	DS	SX		
	RENÉ 150	12% Co	6% Co	0% Co
Ni	BAL	BAL	BAL	BAL
Co	12	12	6	
Cr	5	5	5	5
Al	5.5	5.5	5.5	5.5
Ta	6	6	6	6
V	2.2	2.2	2.2	2.2
Re	3	3	3	3
W	5	5	5	5
Mo	1	1	1	1
Hf	1.5			
Zr	0.03			
C	0.06			
B	0.015			

EFFECT OF COBALT ON ULTIMATE TENSILE STRENGTH



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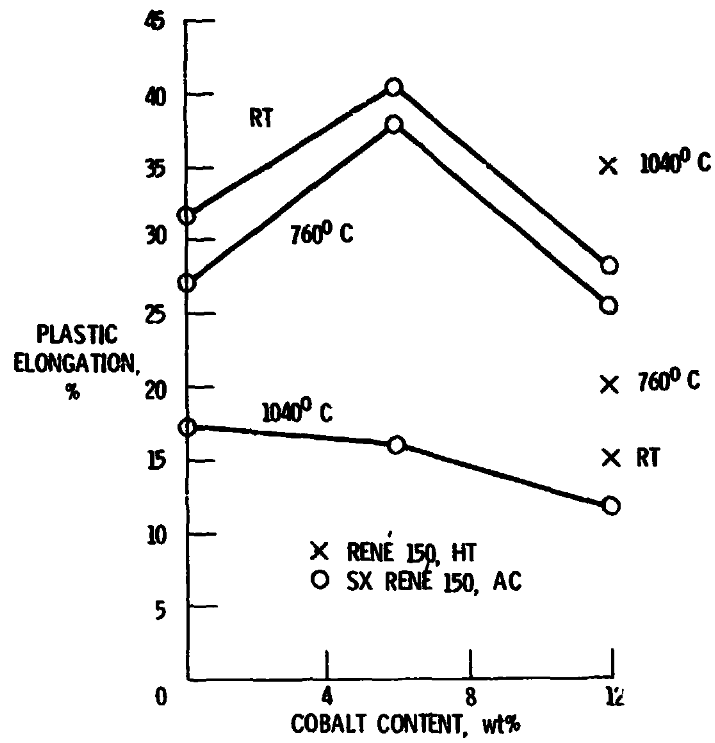
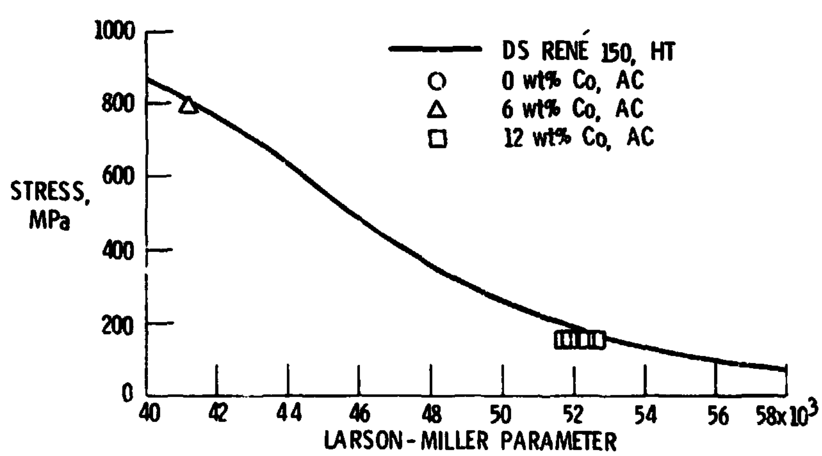
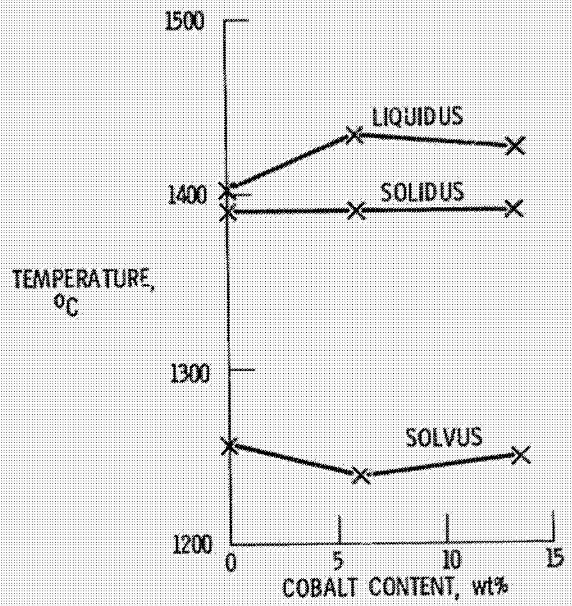
EFFECT OF COBALT ON DUCTILITY



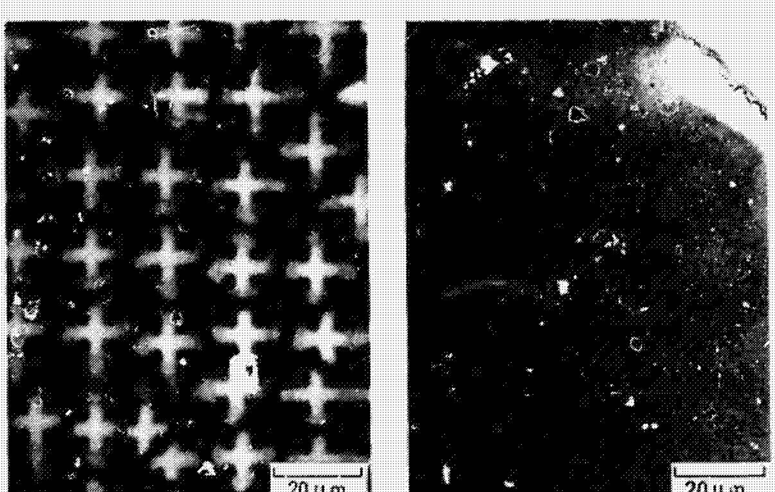
RUPTURE STRENGTH



DTA RESULTS - MOD R-150



TYPICAL MICROSTRUCTURES



AS CAST　　　　1355C/8 hr

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TYPICAL MICROSTRUCTURES

1355C/8 hr

1355C/8 hr
1080C/8 hr

CS-82-2323

RENÉ 150 HEAT TREATMENT

DS	[illegible]
1/2 hr, 1205° C (2200° F)	4 hr, 1355° C (2475° F)
4 hr, 1080° C (1975° F	4 hr, 1080° C (1975° F)
16 hr, 900° C (1650° F)	16 hr, 900° C (1650° F)

TENTATIVE CONCLUSIONS

- DECREASING Co DECREASES RT STRENGTH
- WIDE SCATTER IN 760° C UTS COULD BE DUE TO EITHER
  - —ORIENTATION
  - —Co CONTENT
- UTS AT 1040° C NEARLY CONSTANT
- DUCTILITY DECREASES WITH INCREASING TEMPERATURES IN CONTRAST TO DS RENÉ 150
- 1040° C DUCTILITY INCREASES WITH DECREASING Co
- BELOW 1040° C DUCTILITY HAS MAXIMUM NEAR 6 wt% Co
- SX RENÉ 150 HAS BROAD HT CAPABILITY
- HT AND ORIENTATION IMPORTANT

PLANS

- COMPLETE AC ANALYSIS
  - —S/R
  - —METALLOGRAPHY
  - —X-RAY
- TEST HT SPECIMENS
  - —TENSILE
  - —S/R
- ANALYZE RESULTS
  - —TENSILE
  - —S/R
  - —METALLOGRAPHY
  - —X-RAY

THERMAL FATIGUE RESISTANCE OF COBALT-MODIFIED UDIMET 700

Peter T. Bizon
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

The comparative thermal fatigue resistances of five cobalt composition modifications of Udimet 700 are being determined from fluidized bed tests. Cobalt compositional levels of <0.1, 4.3, 8.6, 12.8, and 17.0 percent are being investigated in both the bare and coated (NiCrAlY overlay) conditions. Triplicate tests of each variation including duplicate tests of three control alloys are being studied. Fluidized beds were maintained at 550° and 1850° F for the first 5500 cycles at which time the hot bed was increased to 1922° F. Immersion time in each bed is always 3 minutes. Upon the completion of 10 000 cycles, it appears that the 8.6 percent cobalt level gives the best thermal fatigue life. Considerable deformation of the test bars is occurring.

WHAT IS THERMAL FATIGUE?

THERMAL FATIGUE IS DEFINED AS THE CRACKING OF A MATERIAL BY ALTERNATE HEATING AND COOLING DURING WHICH FREE THERMAL EXPANSION IS CONSTRAINED. INTERNAL CONSTRAINTS OF AN ELEMENT OF MATERIAL ARE PROVIDED BY ADJACENT MATERIAL ELEMENTS AT A DIFFERENT TEMPERATURE. CONSTRAINT OF THE THERMAL EXPANSION INDUCES THERMAL STRAINS WHICH MAY CAUSE THERMAL FATIGUE CRACKING.

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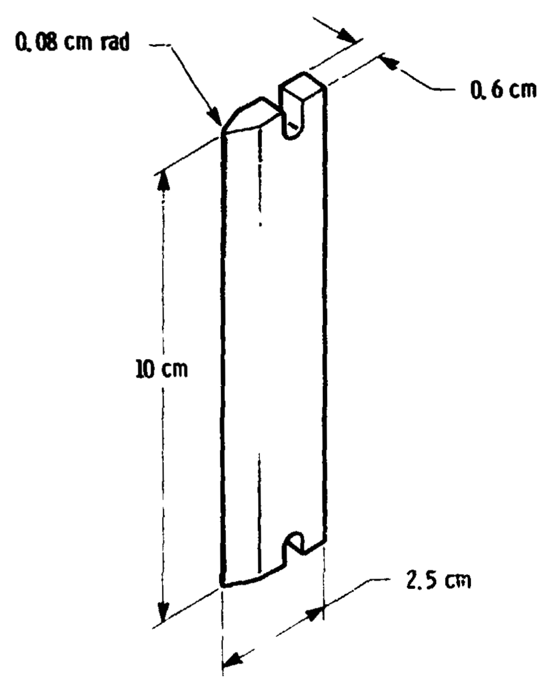
THERMAL FATIGUE SPECIMEN
(SINGLE EDGE WEDGE)



SPECIMENS IN HOLDER

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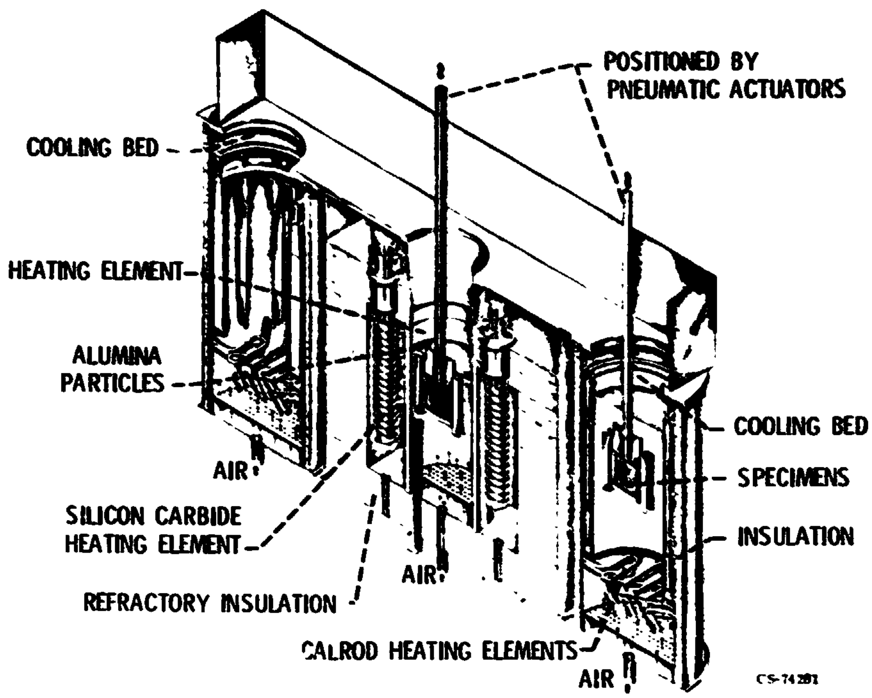
FLUIDIZED BED TEST FACILITY



TEST MATERIALS

NUMBER OF SPECIMENS	ALLOY	% COBALT	BARE/COATED
3	MODIFIED UDIMET 700	17.0	BARE
3	↑	12.8	↑
3		8.6	
3		4.3	↓
3		0.1	BARE
3		17.0	NiCrAlY OVERLAY COATED
3		12.8	↑
3		8.6	
3	↓	4.3	↓
3	MODIFIED UDIMET 700	0.1	NiCrAlY OVERLAY COATED
2	HS 188	—	BARE
2	IN 625	—	BARE
2	IN 800	—	BARE

36 TOTAL SPECIMENS

TEST CONDITIONS

CYCLES	COLD BEDS	HOT BED	IMMERSION TIME IN EACH BED
0 TO 5500	550° F	1850° F	3 min
5500 →	550° F	1922° F	3 min

SUMMARY

- FIVE COBALT VARIATIONS OF UDIMET 700, BOTH BARE AND COATED
- EVALUATED BY SIMULTANEOUS TESTING IN FLUIDIZED BEDS
- APPEARS 8.6% COBALT COMPOSITION GIVES BEST LIFE
- CONSIDERABLE DEFORMATION OCCURRING

N83 11290 D8-26

CREEP-FATIGUE OF LOW COBALT SUPERALLOYS

Gary R. Halford
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

A contractual program has been initiated with the Battelle Columbus Laboratories (BCL) to evaluate the low-cycle fatigue and creep-fatigue resistance of superalloys containing reduced amounts of cobalt. Because of the limited amount of material available for the evaluation, a minimum test matrix was established whereby the creep-fatigue resistance of each composition could be bracketed as quickly as possible using as few specimens as possible. Should the lack of cobalt be determined to significantly alter an alloy's creep-fatigue resistance, then a more comprehensive test matrix could be pursued at a future time for the specific composition(s) of more direct interest.

The test matrix employed at BCL involves a single high temperature appropriate for each alloy. A single total strainrange, again appropriate to each alloy, is used in conducting strain-controlled, low-cycle, creep-fatigue tests. The total strainrange is based upon the level of straining that results in about 10 000 cycles to failure in a high-frequency (0.5 Hz) continuous strain-cycling fatigue test. No creep is expected to occur in such a test. To bracket the influence of creep on the cyclic strain resistance, strain-hold time tests with 1-minute hold periods are introduced into otherwise continuous strain-cycling tests. One test per composition is conducted with the hold period in tension only, one in compression only, and one in both tension and compression. From the results of such a test matrix, it is expected that we will be able to identify those compositions that are prone to significant creep-fatigue interaction. Once identified, the creep-fatigue properties of those compositions could be investigated in greater detail provided the other material properties warrant the effort. The test temperatures, alloys, and their cobalt compositions that are currently under study are given in the attached figures.

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CREEP-FATIGUE OF LOW COBALT SUPERALLOYS

OBJECTIVE:
- DETERMINE EFFECT ON CREEP-FATIGUE RESISTANCE OF COBALT REDUCTION IN SUPERALLOYS

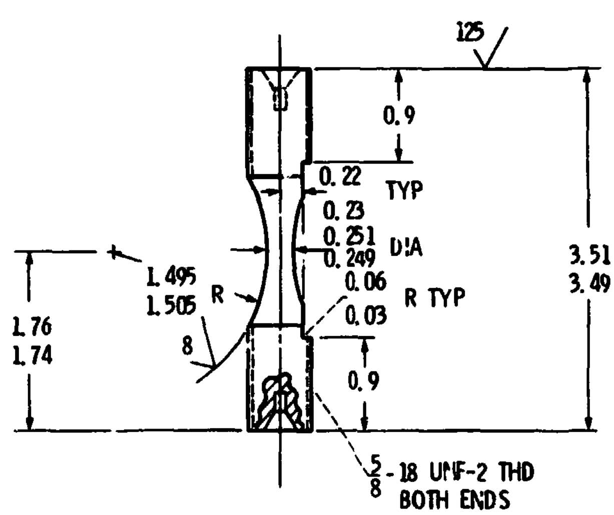
APPROACH:
- EXPERIMENTALLY EVALUATE CREEP-FATIGUE RESISTANCE USING MINIMUM TEST MATRIX TO BRACKET BEHAVIOR
- EXPAND TEST MATRIX IN FUTURE FOR SELECTED COMPOSITIONS

ACTIVITIES:
- PRELIMINARY IN-HOUSE PROGRAM, MARCH-APRIL 1982
- TESTING CONTRACT AWARDED TO BATTELLE COLUMBUS LABORATORIES, MAY 1982 (NAS3-23289)
- EXTEND CONTRACTUAL EFFORT AS NEW MATERIAL BECOMES AVAILABLE

CREEP-FATIGUE OF LOW COBALT SUPERALLOYS

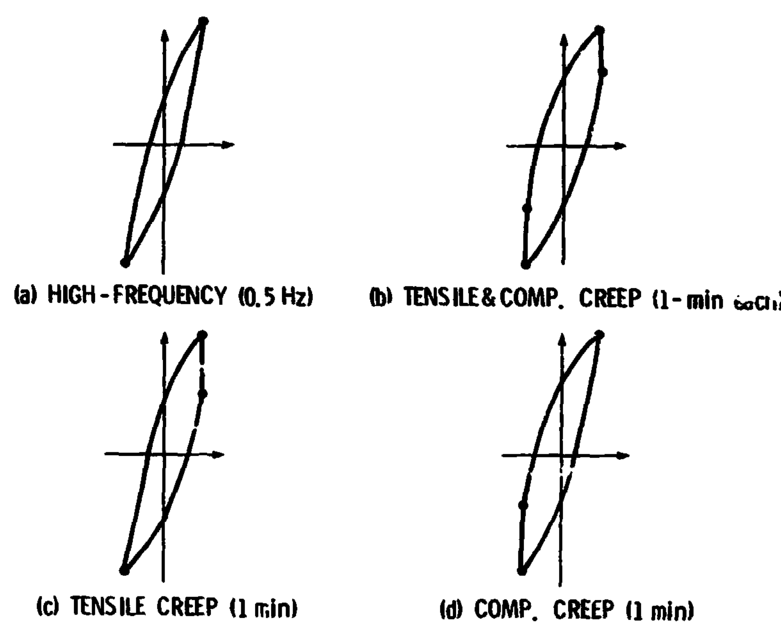
ALLOYS, COBALT COMPOSITIONS, TEST TEMPERATURES, AND SPECIMEN GEOMETRY

ALLOY	COBALT COMPOSITIONS, %					TEST TEMPERATURE, °F
CAST WASPALOY	--	13.5	9.0	4.5	0	1000
WROUGHT U-700	17.0	12.8	8.6	4.3	0	1400
POWDER MET. U-700	17.0	12.8	8.6	4.3	0	1400

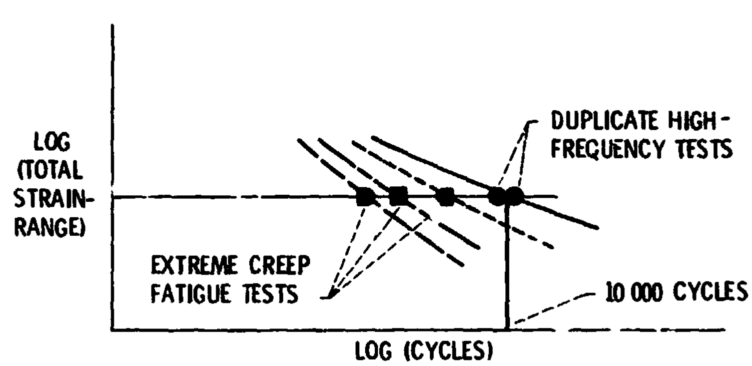


CREEP-FATIGUE OF LOW COBALT SUPERALLOYS

STRAIN-CONTROLLED CREEP-FATIGUE CYCLES



CREEP-FATIGUE OF LOW COBALT SUPERALLOYS



STEP 1. - ESTIMATE BEHAVIOR BASED ON TENSILE AND CREEP PROPERTIES

STEP 2. - SELECT STRAIN RANGE FOR $N_f \sim 10\,000$ HIGH-FREQUENCY CYCLES

STEP 3. - CONDUCT FIRST HIGH-FREQUENCY TEST

STEP 4. - CONDUCT ONE EACH OF EXTREME CREEP-FATIGUE CYCLES

STEP 5. - CONDUCT SECOND HIGH-FREQUENCY TEST OR REPEAT QUESTIONABLE TESTS

STEP 6. - REPEAT TESTING SEQUENCE FOR ALL 19 COMPOSITIONS

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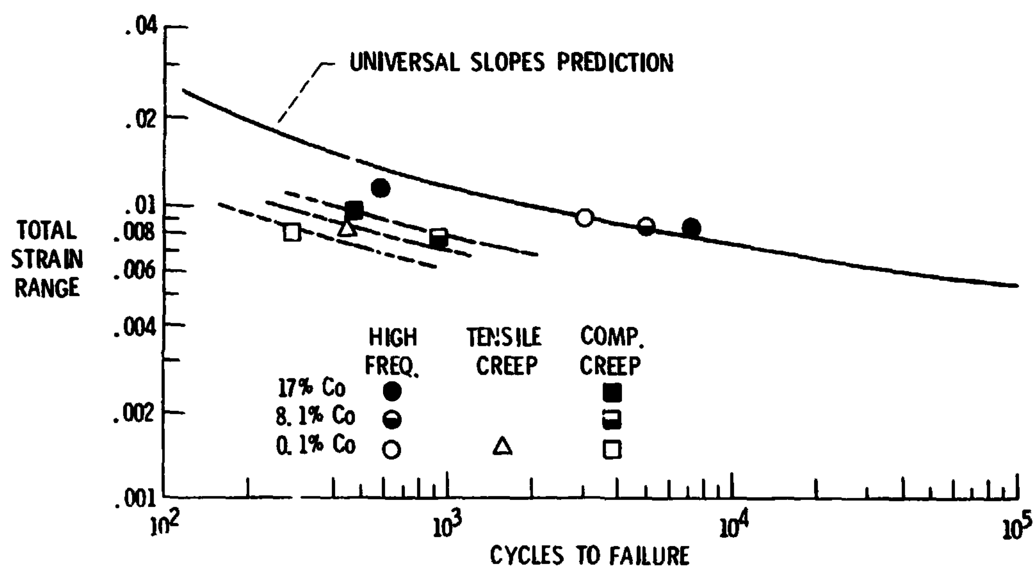
CREEP-FATIGUE OF LOW COBALT SUPERALLOYS

WROUGHT U-700, 1400° F, LeRC RESULTS



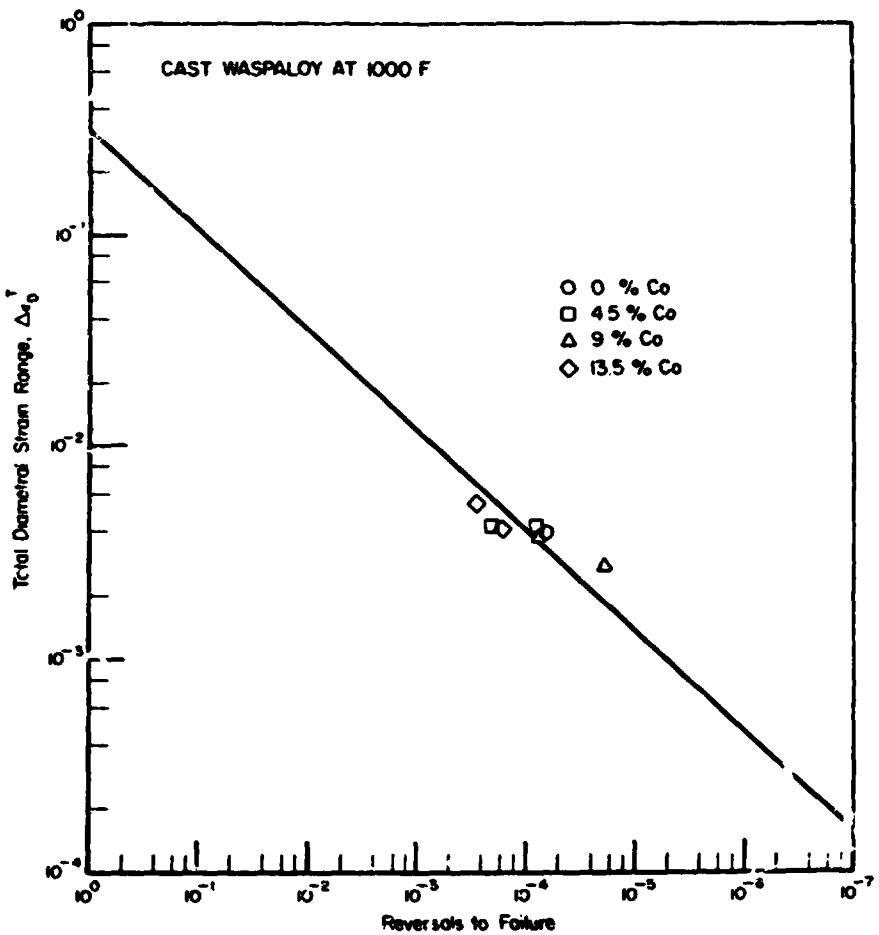


FIGURE 1. CONTINUOUS CYCLING TEST RESULTS ON CAST WASPALOY AT 1000 F AT A FREQUENCY OF 0.5 HZ

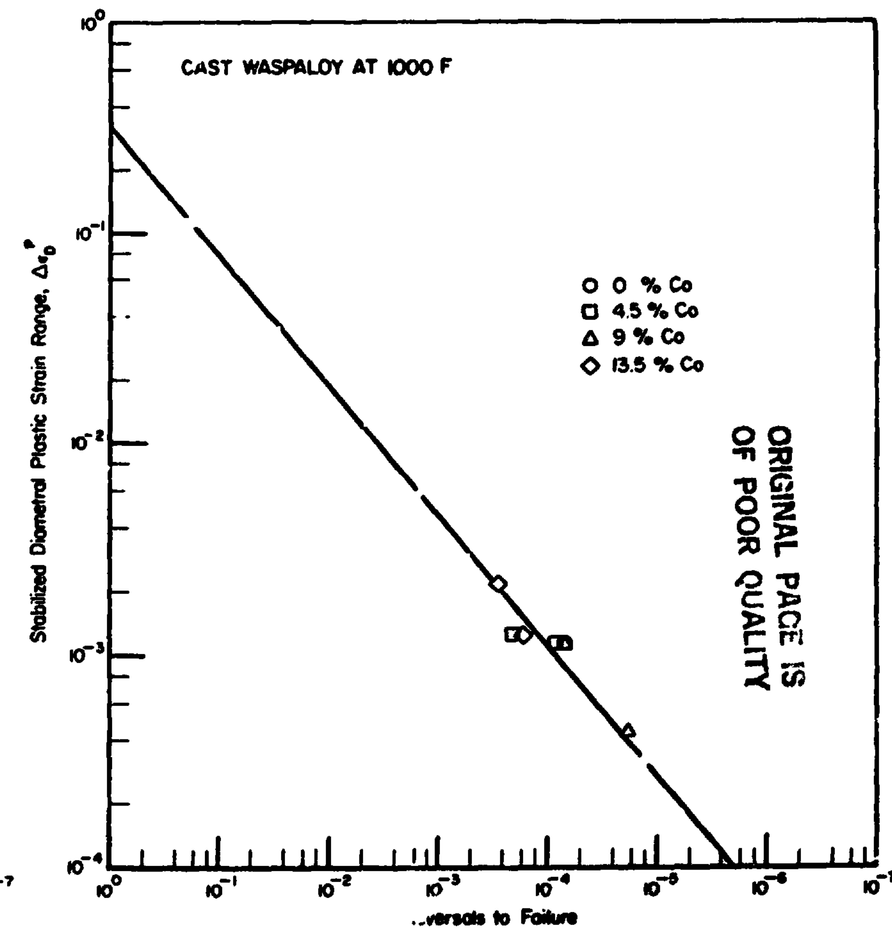


FIGURE 2. CONTINUOUS CYCLING TEST RESULTS ON CAST WASPALOY AT 1000 F AT A FREQUENCY OF 0.5 HZ

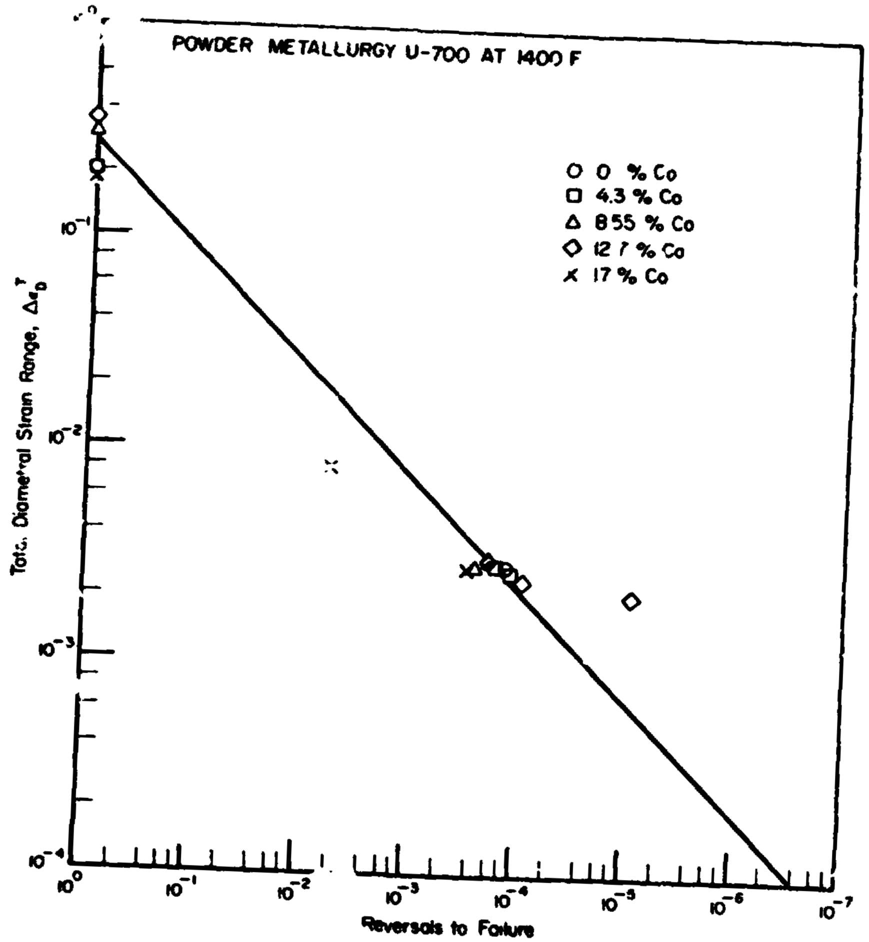


FIGURE 3. CONTINUOUS CYCLING TEST RESULTS ON POWDER METALLURGY U-700 AT 1400 F AT A FREQUENCY OF 0.5 HZ

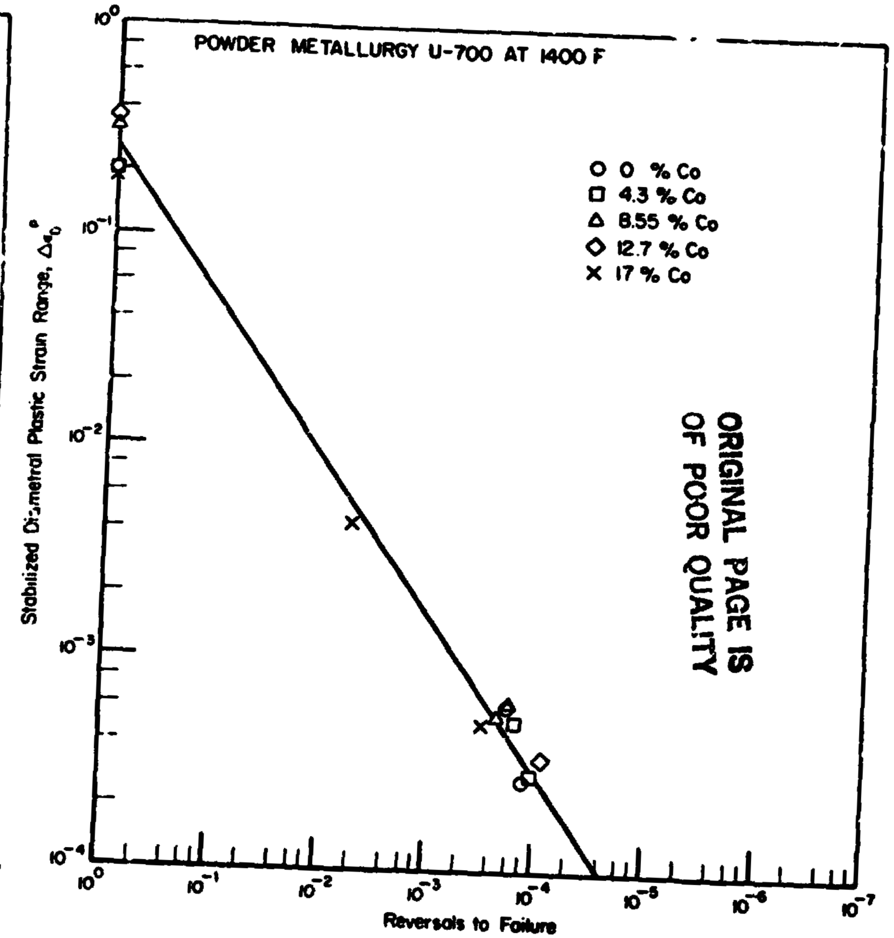


FIGURE 4. CONTINUOUS CYCLING TEST RESULTS ON POWDER METALLURGY U-700 AT 1400 F AT A FREQUENCY OF 0.5 HZ

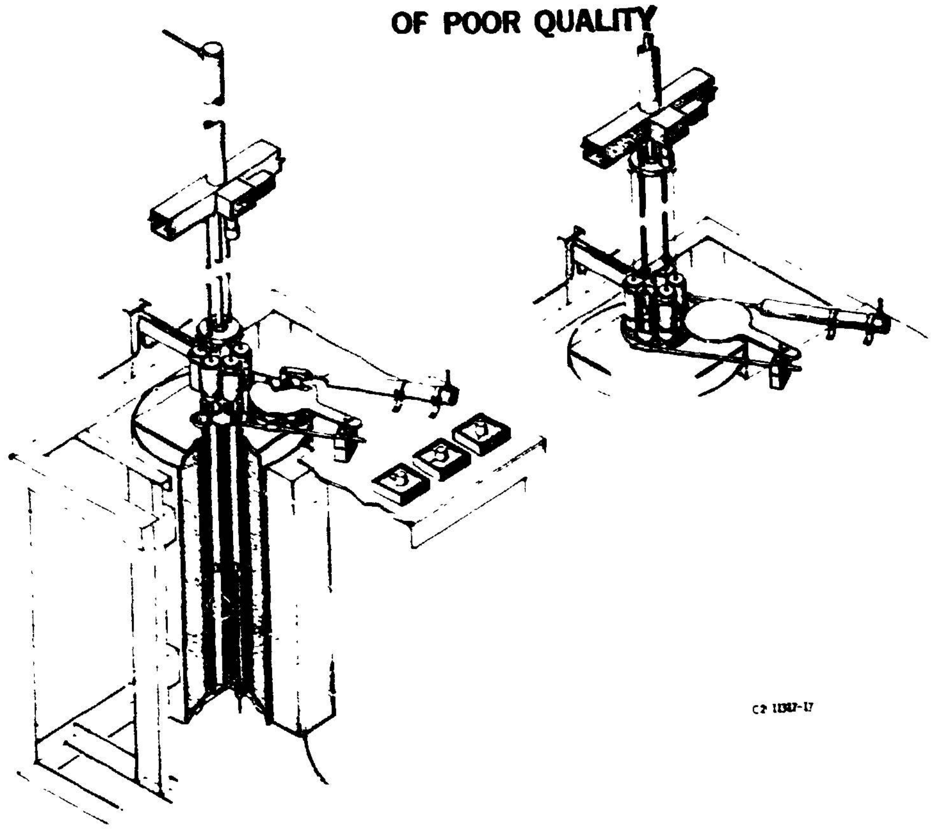
N83 11291 D9

OXIDATION OF LOW COBALT ALLOYS

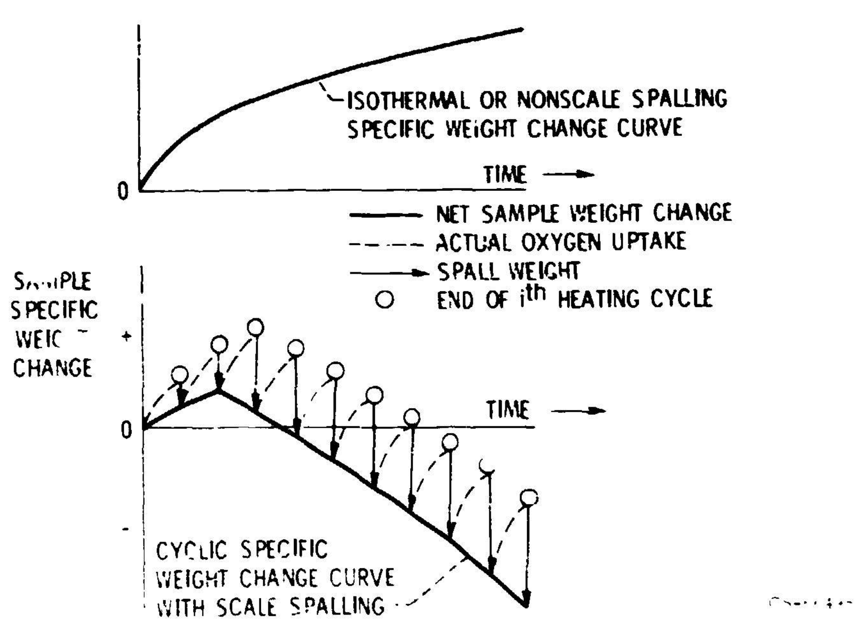
Charles A. Barrett
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

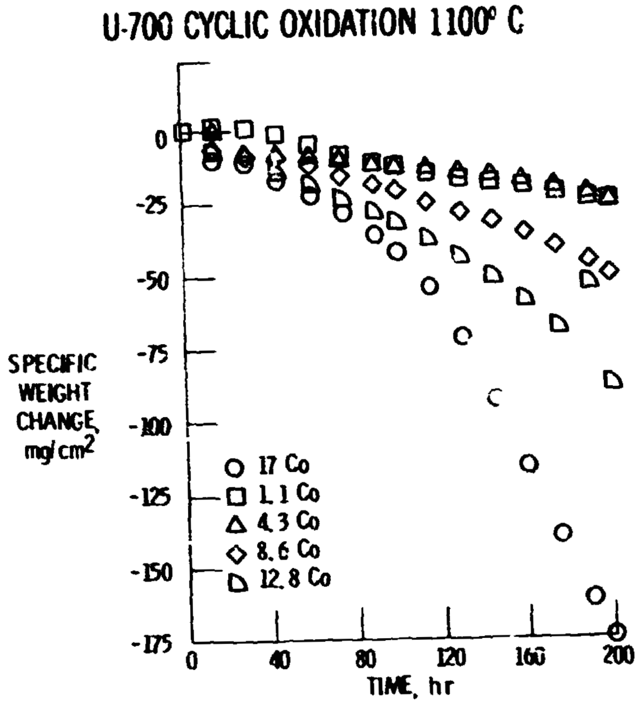
Four high-temperature alloys: U-700, Mar M-247, Waspaloy and PM/HIP U-700 were modified with various cobalt levels ranging from 0 percent to their nominal commercial levels. The alloys were then tested in cyclic oxidation in static air at temperatures ranging from 1000° to 1150° C at times from 500 to 100 1-hour cycles. Specific weight change with time and X-ray diffraction analyses of the oxidized samples were used to evaluate the alloys. The alloys tend to be either Al_2O_3/aluminate spinel or Cr_2O_3/chromite spinel formers depending on the Cr/Al ratio in the alloy. Waspaloy with a ratio of 15:1 is a strong Cr_2O_3 former while this U-700 with a ratio of 3.33:1 tends to form mostly Cr_2O_3 while Mar M-247 with a ratio of 1.53:1 is a strong Al_2O_3 former. The best cyclic oxidation resistance is associated with the Al_2O_3 formers. The cobalt levels appear to have little effect on the oxidation resistance of the Al_2O_3/aluminate spinel formers while any tendency to form Cr_2O_3 is accelerated with increased cobalt levels and leads to increased oxidation attack.

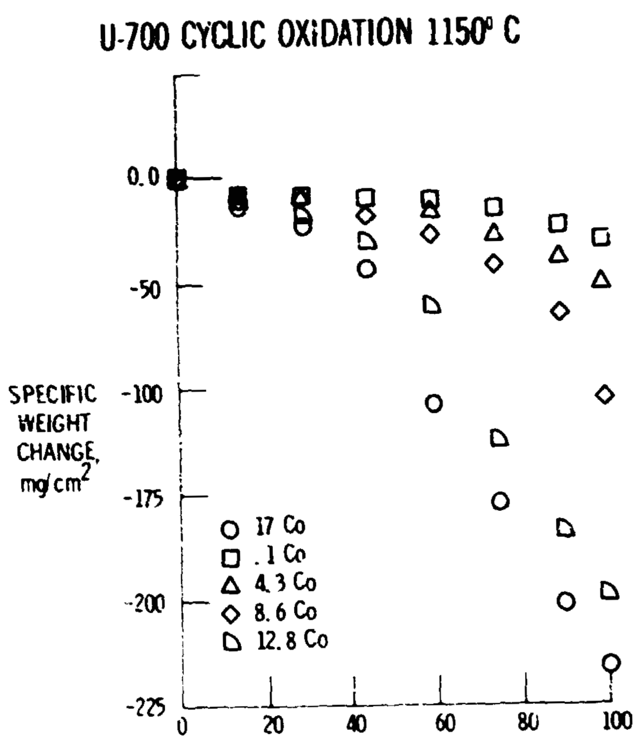
C-2



ISOTHERMAL VS CYCLIC OXIDATION




U-700 CYCLIC OXIDATION 1100° C
0
-25
-50
-75
-100
-125
-150
-175
SPECIFIC WEIGHT CHANGE, mg/cm2
17 Co
1.1 Co
4.3 Co
8.6 Co
12.8 Co
0
40
80
120
160
200
TIME, hr


U-700 CYCLIC OXIDATION 1150° C
0.0
-50
-100
-175
-200
-225
SPECIFIC WEIGHT CHANGE, mg/cm2
17 Co
.1 Co
4.3 Co
8.6 Co
12.8 Co
0
20
40
60
80
100

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FINAL SAMPLE SPECIFIC WEIGHT LOSS

100 1-hr CYCLES 1150^0 C STATIC AIR

	$\Delta W/A$, mg/cm^2
U-700 (17 Co)	-230.72
U-700 12.8 Co	-197.77
U-700 8.6 Co	-104.39
U-700 4.3 Co	-50.07
U-700 .1 Co	-30.50
PM/HIP U-700 (17 Co)	-174.85
PM/HIP U-700 12.7 Co	-142.17
PM/HIP U-700 8.6 Co	-64.70
PM/HIP U-700 4.3 Co	-50.76
PM/HIP U-700 .1 Co	-31.85
WASPALOY (13.5 Co)	-165.20
WASPALLOY 9 Co	-186.13
WASPALLOY 4.5 Co	-103.75
WASPALLOY 0 Co	-94.17
Mar M-247 (9.8 Co)	-19.46
Mar M-247 5.0 Co	-7.95
Mar M-247 .1 Co	-15.26

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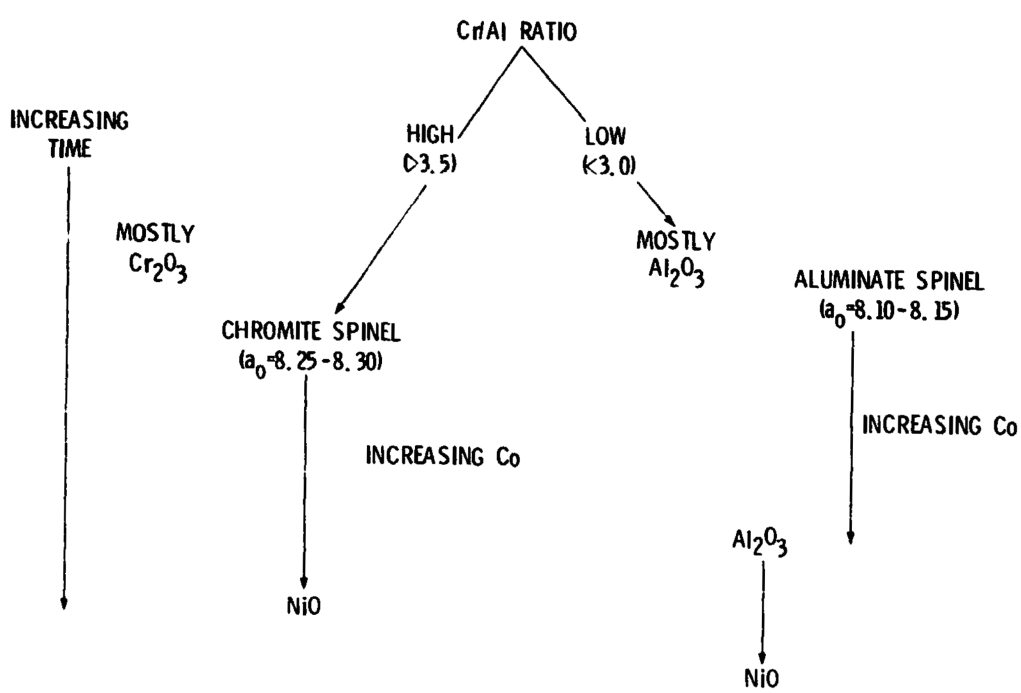
SEQUENCES OF OXIDE FORMATION

100 1-hr CYCLES 1150° C STATIC AIR

	START	END
U-700 (17 Co)	Cr_2O_3, R	NiO, Cr_2O_3, a_0 8.25
U-700 12.8 Co	Cr_2O_3, R	NiO, a_0=8.30, Cr_2O_3
U-700 8.6 Co	Cr_2O_3, R	NiO, a_0=8.30, $NiTiO_2$
U-700 4.3 Co	Cr_2O_3, R	NiO, $NiTiO_3$, a_0=8.25
U-700 .1 Co	Cr_2O_3, R	NiO, $NiTiO_3$, Cr_2O_3
PMHIP U-700 (17 Co)	Cr_2O_3, R, $Cr_xTi_yO_z$	NiO, $NiTiO_3$, Cr_2O_3
PMHIP U-700 12.7 Co	Cr_2O_3, R, $Cr_xTi_yO_z$	NiO, $NiTiO_3$, Cr_2O_3
PMHIP U-700 8.6 Co	Cr_2O_3, R, $Cr_xTi_yO_z$	NiO, $NiTiO_3$, Cr_2O_3
PMHIP U-700 4.3 Co	Cr_2O_3, R, $Cr_xTi_yO_z$	NiO, a_0=8.25, Cr_2O_3
PMHIP U-700 .1 Co	Cr_2O_3, R, $Cr_xTi_yO_z$	NiO, a_0=8.25, Cr_2O_3
WASPALOY (13.5 Co)	Cr_2O_3, R, NiO	NiO, a_0=8.30, Cr_2O_3
WASPALOY 9.0 Co	Cr_2O_3, R, NiO	NiO, a_0=8.30, Cr_2O_3
WASPALOY 4.5 Co	Cr_2O_3, R, NiO	Cr_2O_3, NiO, a_0=8.30
WASPALOY 0 Co	Cr_2O_3, R, NiO	Cr_2O_3, NiO, a_0=8.30
Mar M-247 (9.8 Co)	a_0=8.25, R, $Ni(W, Mo)O_4$	a_0=8.10, Al_2O_3, R
Mar M-247 5.0 Co	a_0=8.25, R, $Ni(W, Mo)O_4$	a_0=8.10, Al_2O_3, R
Mar M-247 .1 Co	a_0=8.25, R, $Ni(W, Mo)O_4$	a_0=8.10, Al_2O_3, R

* R= RUTILE



Cr/Al RATIOS		
	WASPALOY	15: 1
	U-700	3.3: 1
	Mar M-247	1.53: 1

ALLOY SCALING TENDENCY IN HIGH TEMPERATURE OXIDATION

HOT CORROSION OF LOW COBALT ALLOYS

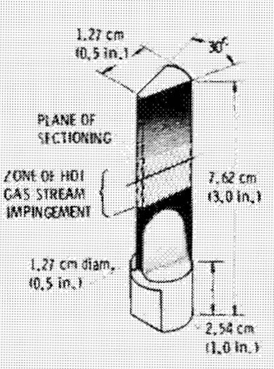
Carl A. Stearns
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

As part of the COSAM program, we have been investigating the hot corrosion attack susceptibility of various alloys as a function of strategic materials content. Preliminary results have been obtained for two commercial alloys, Udimet 700 and Mar-M 247, that were modified by varying the cobalt content. For both alloys the cobalt content was reduced in steps to zero. Nickel content was increased accordingly to make up for the reduced cobalt but all other constituents were held constant. Wedge bar test samples were produced by casting. The hot corrosion test consisted of cyclically exposing samples to the high velocity flow of combustion products from an air-fuel burner fueled with jet A-1 and seeded with a sodium chloride aqueous solution. The flow velocity was Mach 0.5 and the sodium level was maintained at 0.5 ppm in terms of fuel plus air. The test cycle consisted of holding the test samples at 900° C for 1 hour followed by 3 minutes in which the sample could cool to room temperature in an ambient temperature air stream. Assessing the extent of hot corrosion attack has proved to be a challenge and various methods are being evaluated. Every 15 cycles the sample is placed in a coil and the inductance of the coil plus sample combination is measured. This is a nondestructive method, and results to date indicate that change of inductance can be related to extent of attack and useful life. At the end of 200 cycles samples were electrolytically descaled, weighed, mounted and cross sectioned for metallographic examination to ascertain the extent of attack and amount of unattached alloy remaining. For both alloys tested, hot corrosion attack appeared to decrease as the cobalt content was reduced. Final measurements of the attack have not been completed but the preliminary results indicate that cobalt is deleterious with respect to the hot corrosion attack produced by the test method employed. Further evaluation of the role of cobalt is still in progress.

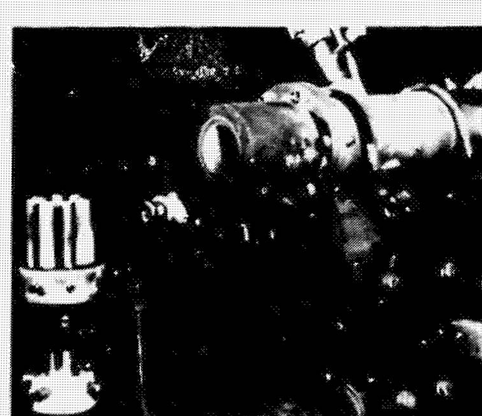
ALLOY CHEMISTRY, IN wt PERCENT, FOR HOT CORROSION TEST SAMPLES

ALLOY	Co	Si	Cr	Ta	Al	Mo	W	Ti	Fe	Hf	Zr	Mn	C
UDIMET 700 - COMMERCIAL	15.5	0.1	14.2	0.0	4.2	4.4	0.0	3.3	0.1	0.0	0.0	0.0	0.1
UDIMET 700 - MODIFICATION 1	17.0	0.1	14.9	0.0	4.1	5.0	0.0	3.6	0.1	0.0	0.0	0.1	0.1
UDIMET 700 - MODIFICATION 2	12.8	0.1	14.7	0.0	4.1	5.0	0.0	3.6	0.1	0.0	0.0	0.1	0.1
UDIMET 700 - MODIFICATION 3	8.6	0.1	15.0	0.0	4.1	5.1	0.0	3.5	0.1	0.0	0.0	0.1	0.1
UDIMET 700 - MODIFICATION 4	4.3	0.1	15.1	0.0	4.1	4.9	0.0	3.6	0.2	0.0	0.0	0.1	0.1
UDIMET 700 - MODIFICATION 5	0.1	0.1	15.1	0.0	4.1	5.0	0.0	3.5	0.1	0.0	0.0	0.1	0.1
Mar M-247 - COMMERCIAL	9.8	0.0	8.4	3.0	5.5	0.7	9.8	1.0	0.1	1.5	0.0	0.0	0.1
Mar M-247 - MODIFICATION 1	5.0	0.0	8.5	3.2	5.4	0.7	10.5	0.9	0.0	1.0	0.1	0.0	0.1
Mar M-247 - MODIFICATION 2	0.1	0.0	8.4	3.9	5.1	0.6	10.2	1.0	0.0	1.0	0.1	0.0	0.1

HOT-CORROSION APPARATUS AND TEST SPECIMEN



TEST BAR



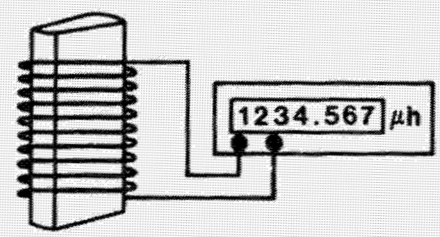
BURNER RIG

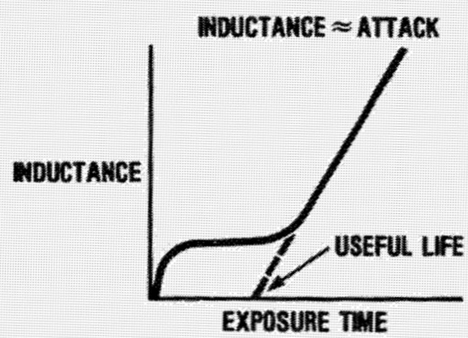
CS-80-1262

NON-DESTRUCTIVE METHOD FOR MEASURING HOT CORROSION OF TURBINE MATERIALS

BURNER RIG OR FURNACE CORROSION EXPOSURE

CORRODED SAMPLE





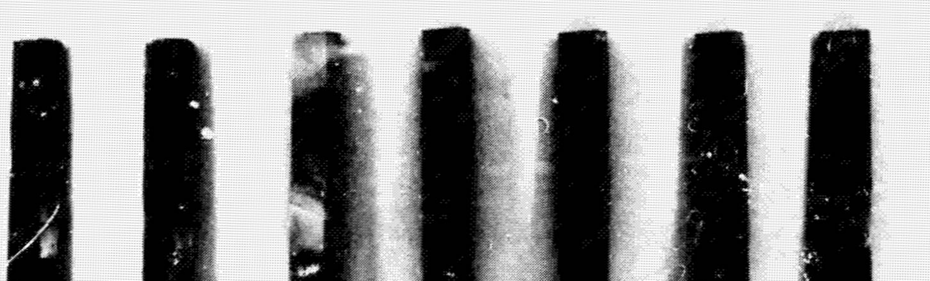
EFFECT OF COBALT ON HOT CORROSION

170 ONE HOUR CYCLES

900 C°

0.5 w PPM Na as NaCl

MACH 0.5



COMMERCIAL U-700
2 VENDORS

OCo 4.3 Co 8.6 Co 12.8 Co 17.0 Co

MODIFIED U-700

CD-82-12950

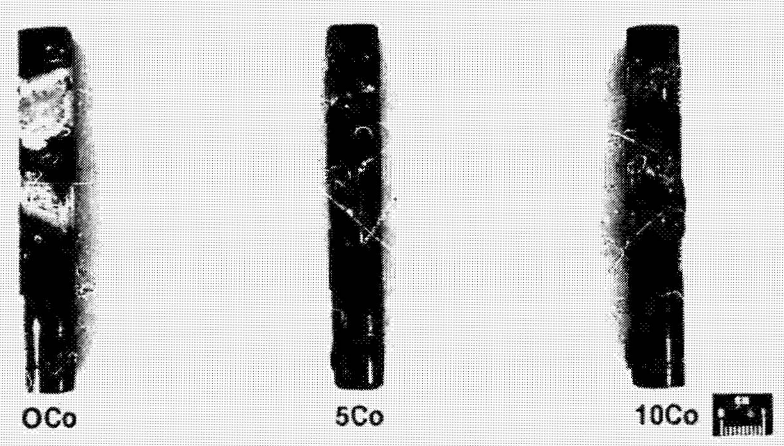
EFFECT OF COBALT ON HOT CORROSION

170 ONE HOUR CYCLES

900 C°

0.5 w PPM Na as NaCl

MACH 0.5



MODIFIED MAR-M 247

CD-82-12951

COATINGS FOR COSAM ALLOYS

Isidor Zaplatynsky and Stanley R. Levine
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

A program to investigate the effects of alloy strategic element content on the burner rig oxidation lives of typical high-temperature metallic coatings has been initiated. The first phase of this effort involves an investigation of the effects of U-700 and Mar-M 247 cobalt-content on the oxidation lives of a typical aluminide coating and a typical low pressure plasma sprayed NiCrAlYSi coating. Early data for the aluminide coated alloys shows an effect of cobalt-content on coating/substrate interdiffusion and on oxidation behavior. The second phase of this effort entails a statistically designed experiment to study the effects of Cr, Al, Co, Ta, and Mo on coating life. Materials for this effort are being prepared.

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COATINGS FOR COSAM ALLOYS

OBJECTIVE: DETERMINE EFFECTS OF ALLOY STRATEGIC METAL CONTENT ON COATING STABILITY AND LIFE

APPROACH:
- DETERMINE EFFECTS OF COBALT/TANTALUM LEVEL IN COSAM A'LOYS (U-700, Mar-M 247, ETC.) ON LIFE OF ALUMINIDE AND ' ERLAY COATINGS
- CONDUCT BROADER INVESTIGATION OF ALLOYING EFFECTS (Cr, Co, Ta, Al, Mo) ON COATING LIFE

EFFECTS OF COSAM ALLOY COBALT/TANTALUM LEVEL ON COATING LIFE

ALLOYS

- U-700 - Co LEVEL
  - WROUGHT - 5 LEVELS
  - CAST - 1 LEVEL
  - PM - 2 LEVELS
- MarM-247 - Co LEVEL
  - CAST - 3 LEVELS
- TANTALUM - ALLOY TBD

COATINGS

- PLASMA SPRAYED NiCoCrAlYSi
- ALUMINIDE

MACH 0.3 BURNER RIG OXIDATION

- ONE TEMPERATURE (1100° C)
- 1-hr CYCLES (TIMES TBD)

EFFECTS OF ALLOY COMPOSITION ON COATING LIFE

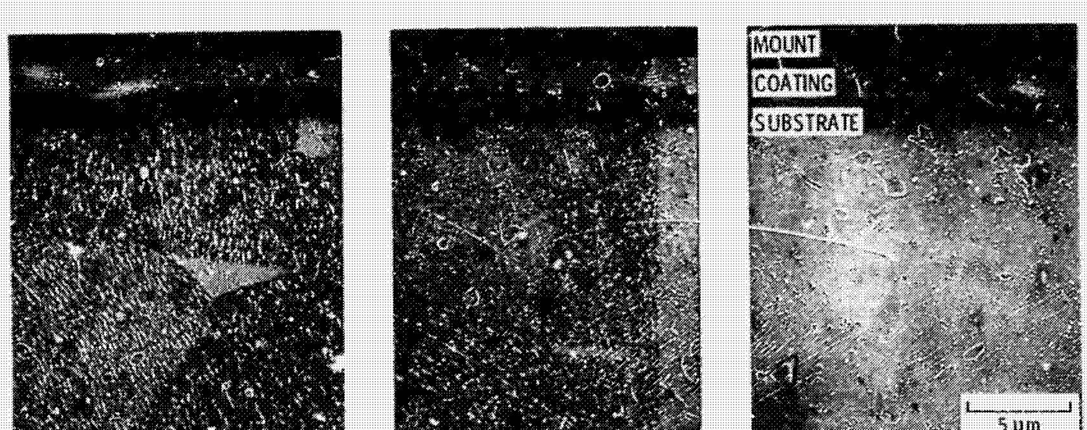
- BASE ALLOY - U-700 (SAME AS IN HOST)
  VARIABLES - Cr, Al, Co, Ta, Mo
  - STATISTICALLY DESIGNED EXPERIMENT

  CHECKS ON REFRACTORY ELEMENT SUBSTITUTION
  W -- Mo
  Nb -- Ta
- COATINGS

  PLASMA SPRAYED NiCoCrAlYSi
  ALUMINIDE (SAME AS IN HOST)
- EVALUATION

  MACH 0.3 BURNER RIG OXIDATION
  1100^0 C, 1-hr CYCLES, DURATION TBD

EFFECT OF COBALT ON ALUMINIZATION OF U-700

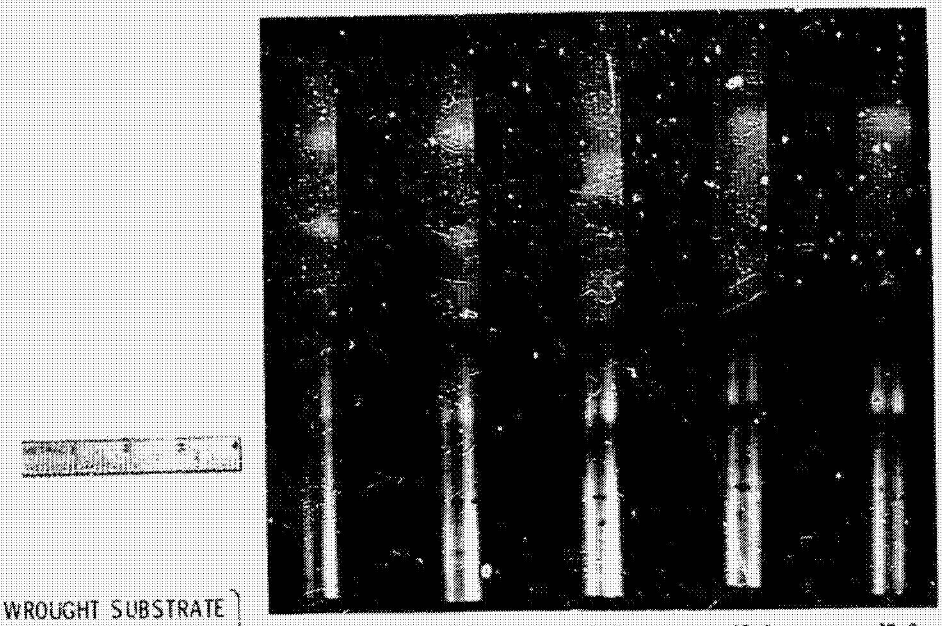


WROUGHT SUBSTRATE COBALT CONTENT, wt%	0.1	8.6	17.0

CS-80-2247

EFFECT OF COBALT CONTENT ON OXIDATION OF ALUMINIZED U-700

MACH 0.3 BURNER RIG, 150 1-hr CYCLES FRONT FACE: 1100° C

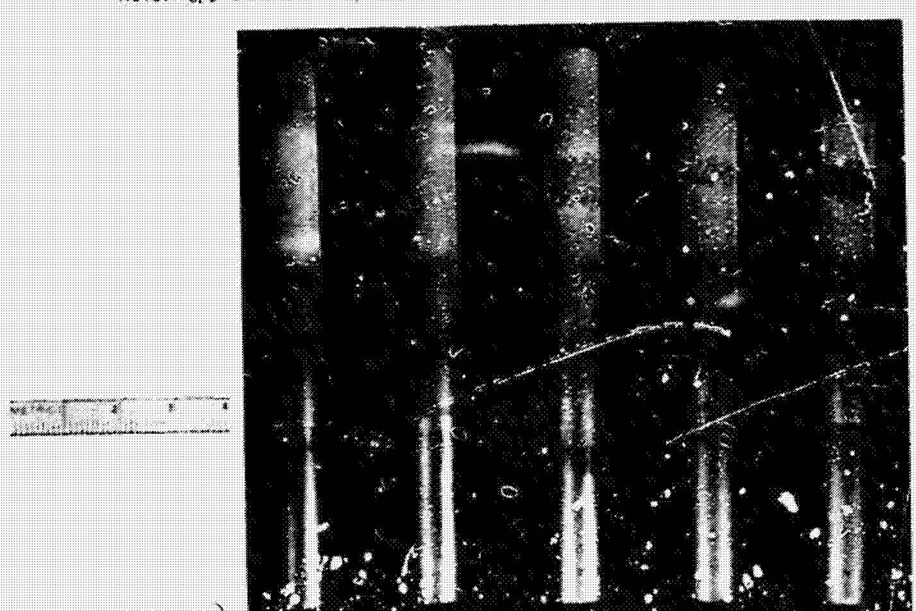


WROUGHT SUBSTRATE COBALT CONTENT, wt%: 0.1 | 4.3 | 8.6 | 12.8 | 17.0

CS-82-2262

EFFECT OF COBALT CONTENT ON OXIDATION OF ALUMINIZED U-700

MACH 0.3 BURNER RIG, 150 1-hr CYCLES BACK FACE: 1120° C

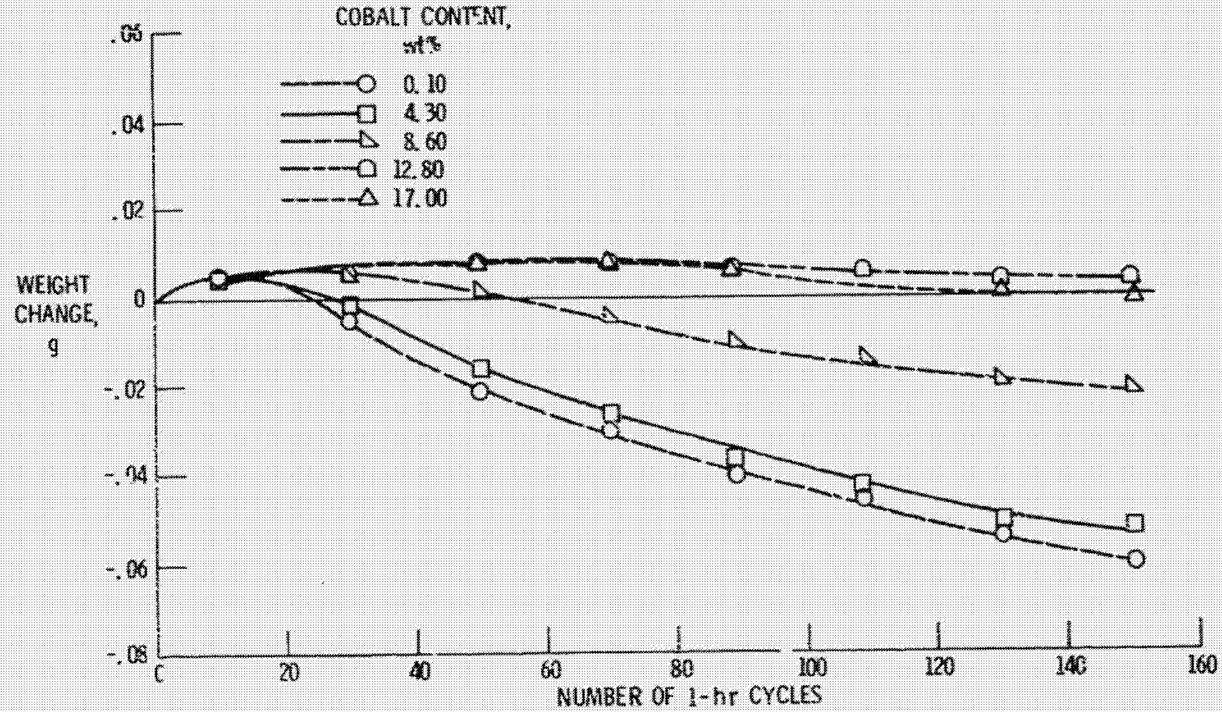


WROUGHT SUBSTRATE COBALT CONTENT, wt%: 0.1 | 4.3 | 8.6 | 12.8 | 17.0

CS-82-2250

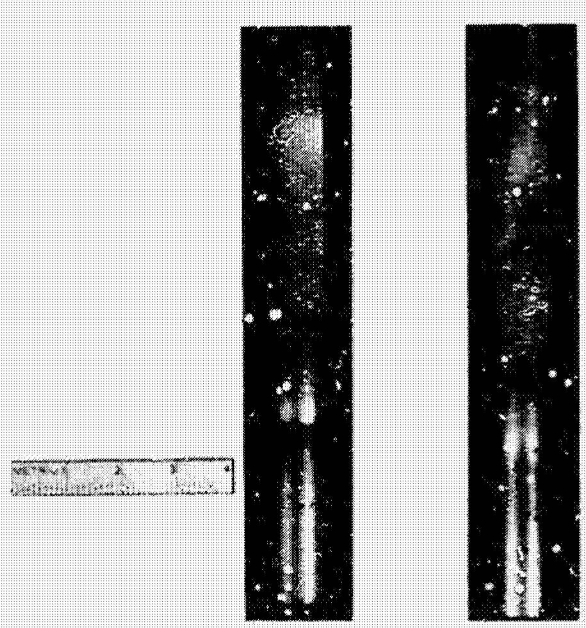
EFFECT OF COBALT ON OXIDATION BEHAVIOR OF ALUMINIZED U-700

MACH 0.3 BURNER RIG, 1100° C (FRONT FACE)



OXIDATION OF ALUMINIDE ON WROUGHT AND CAST U-700 ALLOYS

MACH 0.3 BURNER RIG, 150 1-hr CYCLES FRONT FACE: 1100° C

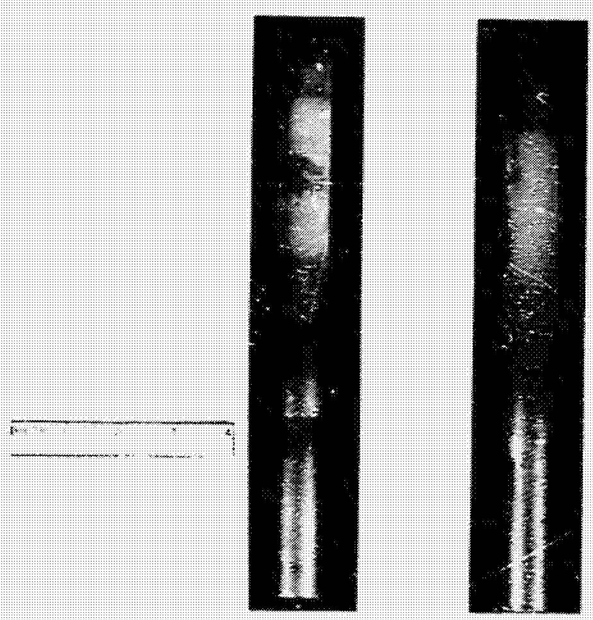


SUBSTRATE	WROUGHT	CAST
COBALT CONTENT, wt%	17.0	19.0

CS-82-2263

OXIDATION OF ALUMINIDE ON WROUGHT AND CAST U-700 ALLOYS

MACH 0.3 BURNER RIG, 150 1-hr CYCLES BACK FACE: 1120° C

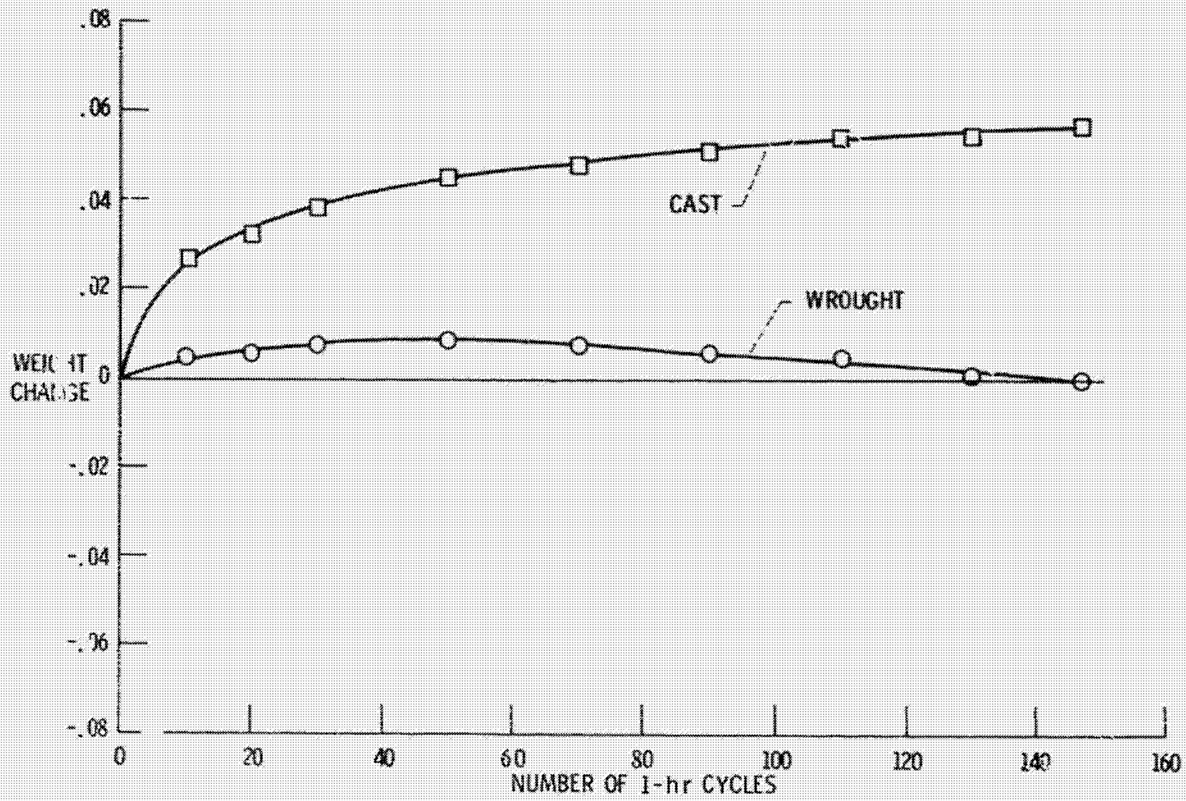


SUBSTRATE	WROUGHT	CAST
COBALT CONTENT, wt%	17.0	19.0

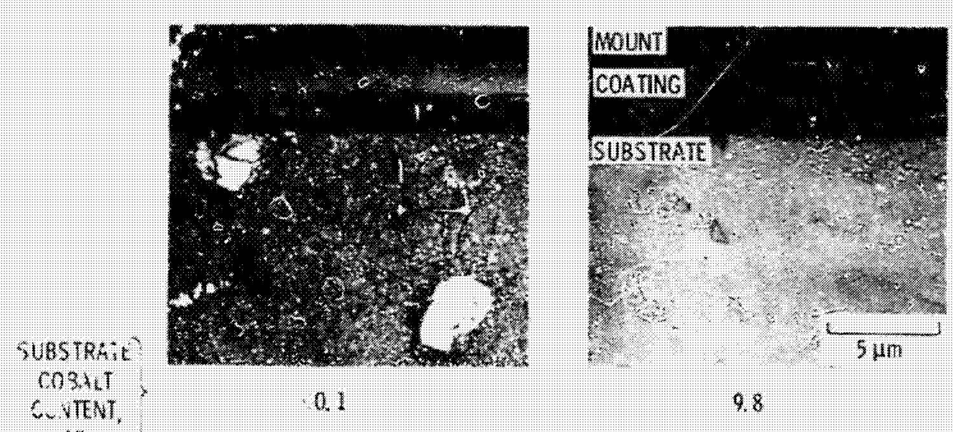
CS-82-2249

OXIDATION BEHAVIOR OF ALUMINIZED U-700 - 18 wt% COBALT

MACH 0.3 BURNER RIG, 1100° C (FRONT FACE)



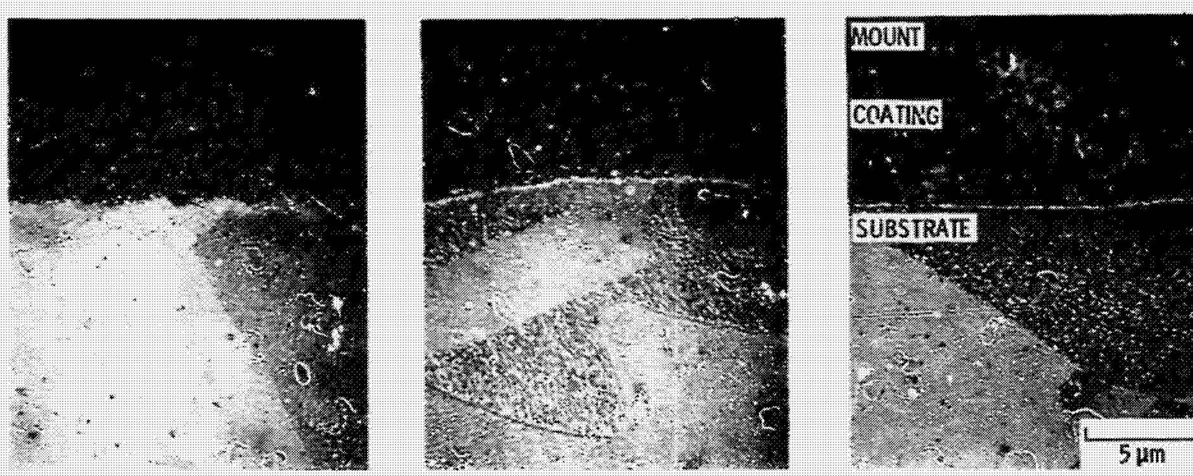
EFFECT OF Co ON ALUMINIZATION OF MarM-247



CS-82-2245

EFFECTS OF U-700 COBALT ON STRUCTURE OF HT NiCrAlYSi

4 hr, 1080° C, ARGON



SUBSTRATE COBALT CONTENT, wt%: 0.1 | 8.6 | 17.0

CS-82-2248

EFFECTS OF MarM 247 COBALT ON STRUCTURE OF HT NiCrAlYSi

4 hr, 1080° C, ARGON

SUBSTRATE COBALT CONTENT, wt%: 0.1 | 9.8

CS-82-2246

N83 11293 D11

INFLUENCE OF COBALT, TANTALUM, AND TUNGSTEN ON THE MICROSTRUCTURE AND MECHANICAL PROPERTIES OF SUPERALLOY SINGLE CRYSTALS

Michael V. Nathal
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

and

L. J. Ebert
Case Western Reserve University
Cleveland, Ohio

The purpose of this study was to investigate the influence of Co, Ta, and W on the microstructure and mechanical properties of nickel-base superalloy single crystals. A matrix of alloys was based on Mar-M 247 stripped of C, B, Zr, and Hf. The microstructures of the alloys were examined using optical and electron microscopy, phase extraction, X-ray diffraction, and differential thermal analysis. Tensile and creep-rupture tests were performed at 1000° C. An increase in tensile and creep strength resulted when Co was removed from alloys containing high refractory metal contents, but Co effects were negligible for alloys with lower refractory metal levels. In the composition range studied, W was more effective than Ta in increasing the creep resistance. The mechanical properties will be discussed in relation to the microstructures of the alloys.

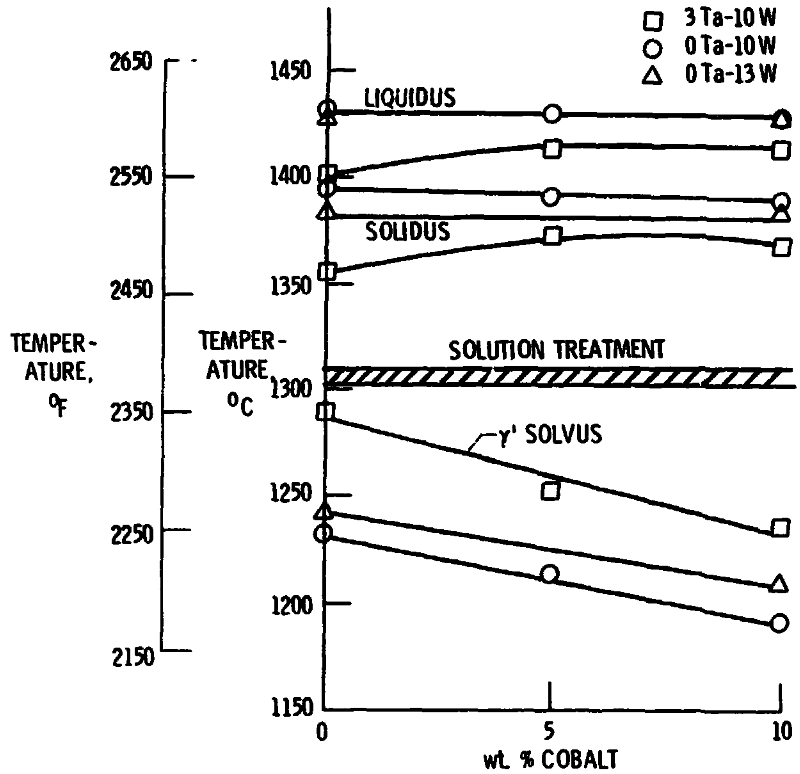
- COSAM
- TENSILE AND CREEP-RUPTURE TESTS AT 1000° C
- MICROSTRUCTURAL FEATURES: γ' VOLUME FRACTION
  γ' COARSENING RATE
  γ,γ' COMPOSITION
  γ-γ' MISMATCH
  TCP FORMATION

SINGLE CRYSTAL ALLOY MATRIX

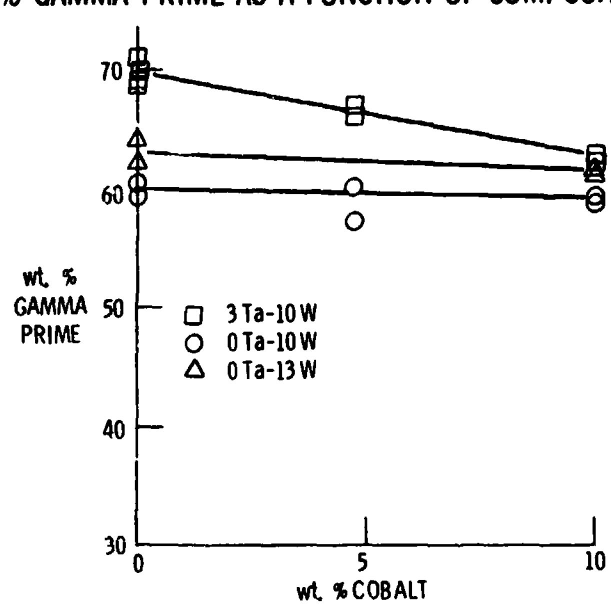
ALLOY	COMPOSITION			TCP FORMATION*	NOTES
	Co	Ta	W		
A	0	0	10	NONE	
B	0	3	10	1 TO 2%	NASAIR 100 2 CASTINGS
C	0	0	13	<1%	
D	5	0	10	NONE	
E	5	3	10	NONE	~ALLOY 3 (-Hf)
F	10	0	10	NONE	
G	10	3	10	NONE	"STRIPPED" Mar-M247
H	10	0	13	NONE	

*SOLUTION TREATED PLUS 1000 hr AT 1000° C

TRANSFORMATION TEMPERATURES AS A FUNCTION OF COMPOSITION



wt. % GAMMA PRIME AS A FUNCTION OF COMPOSITION



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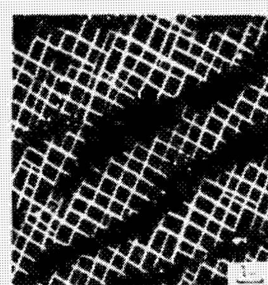
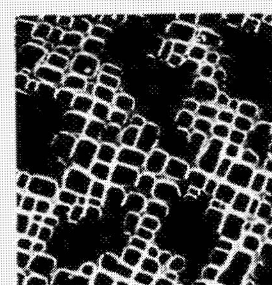
MICROSTRUCTURES OF SELECTED ALLOYS

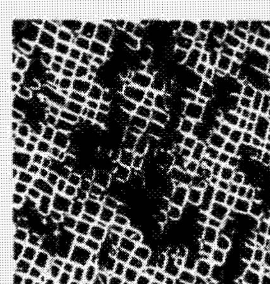
Solution treated plus 100 hrs. at 1000°C



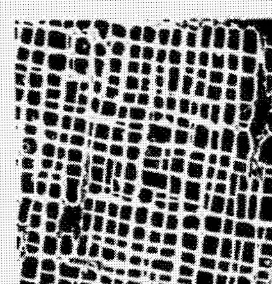
0Co 3Ta 10W



0Co 0Ta 10W

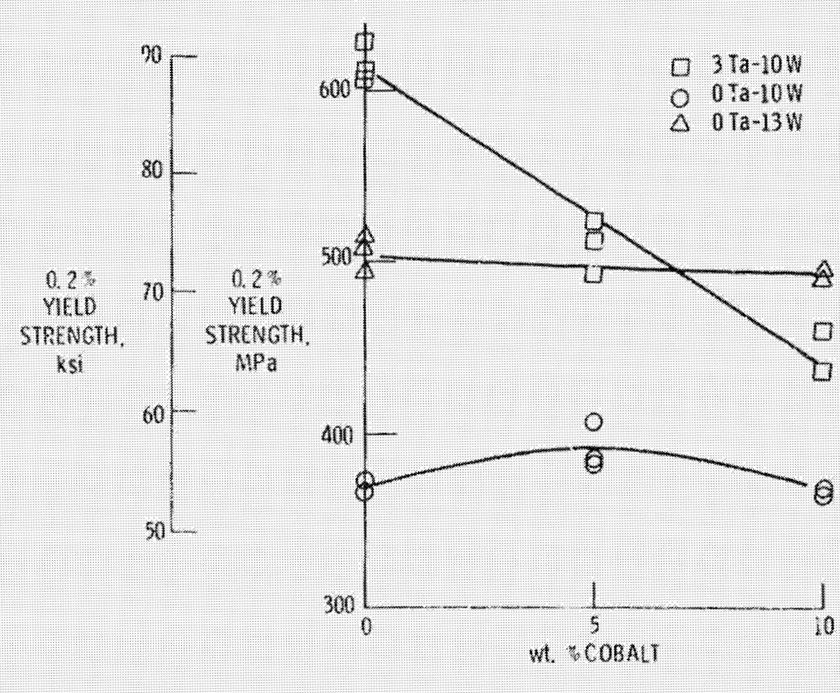


10Co 3Ta 10W



0Co 0Ta .3W

1000° C YIELD STRENGTH OF [100] ORIENTED SINGLE CRYSTALS AS A FUNCTION OF COMPOSITION



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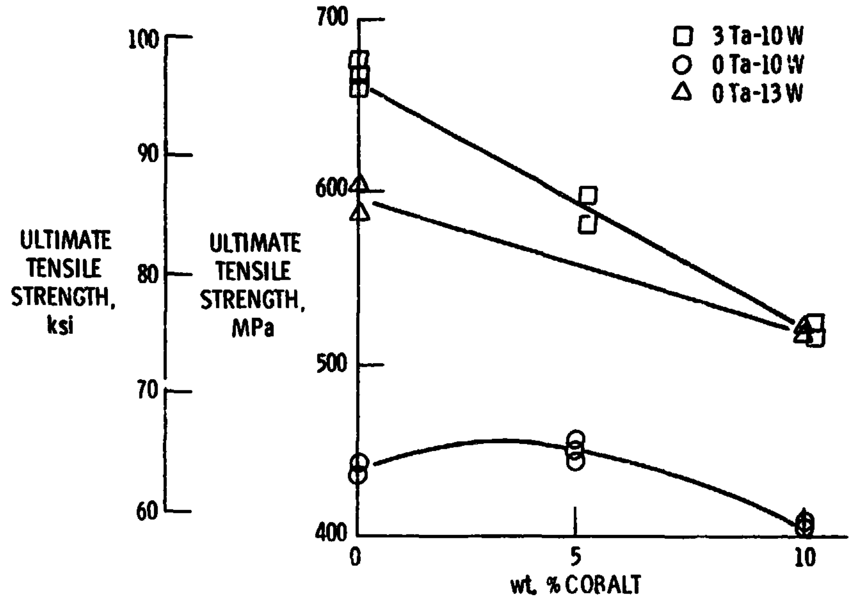
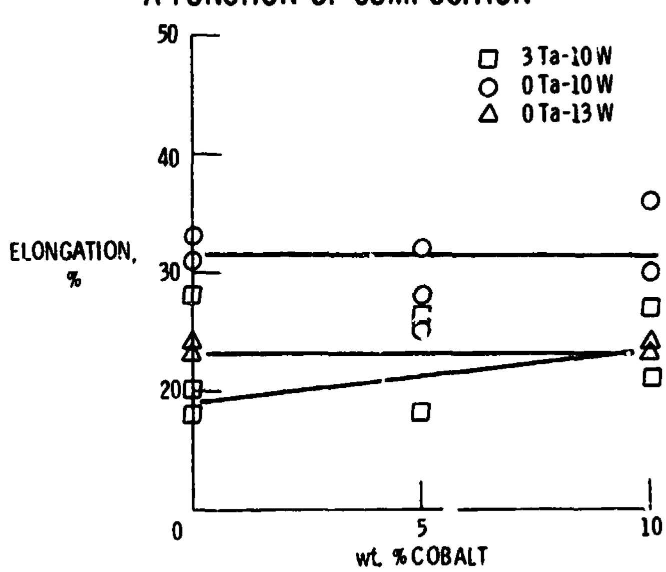
1000° C ULTIMATE TENSILE STRENGTH OF [100] ORIENTED SINGLE CRYSTALS AS A FUNCTION OF COMPOSITION



1000° C TENSILE DUCTILITY OF [100] ORIENTED SINGLE CRYSTALS AS A FUNCTION OF COMPOSITION



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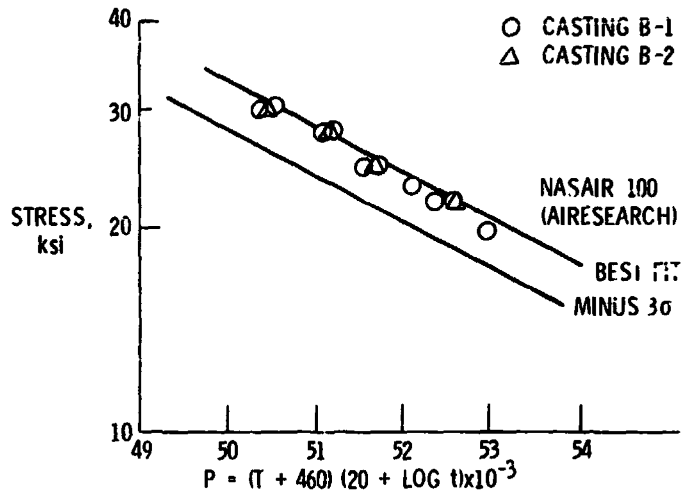
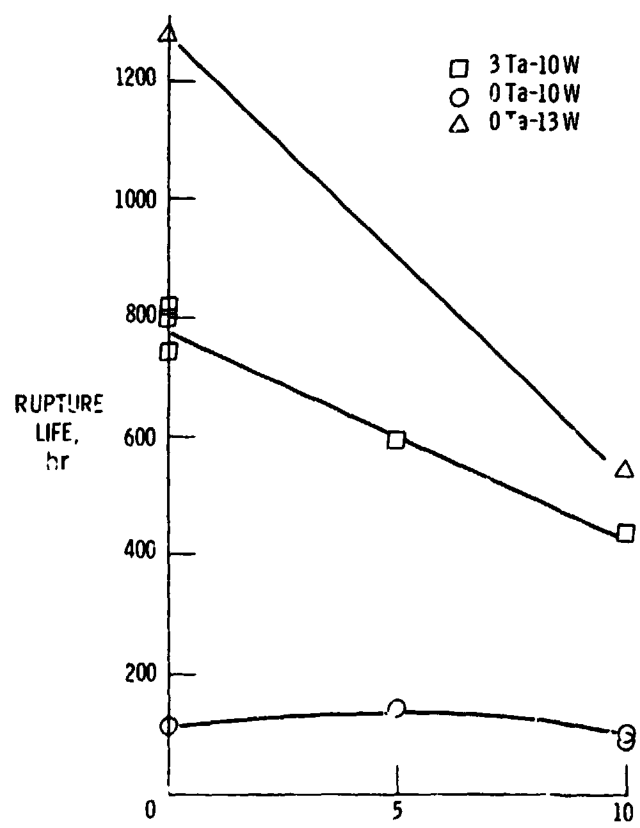
STRESS RUPTURE LIFE OF ALLOY B (0 Co-3 Ta-10 W)



RUPTURE LIFE OF [100] ORIENTED SINGLE CRYSTALS AS A FUNCTION OF COMPOSITION

1000° C, 148 MPa (21.5 ksi)



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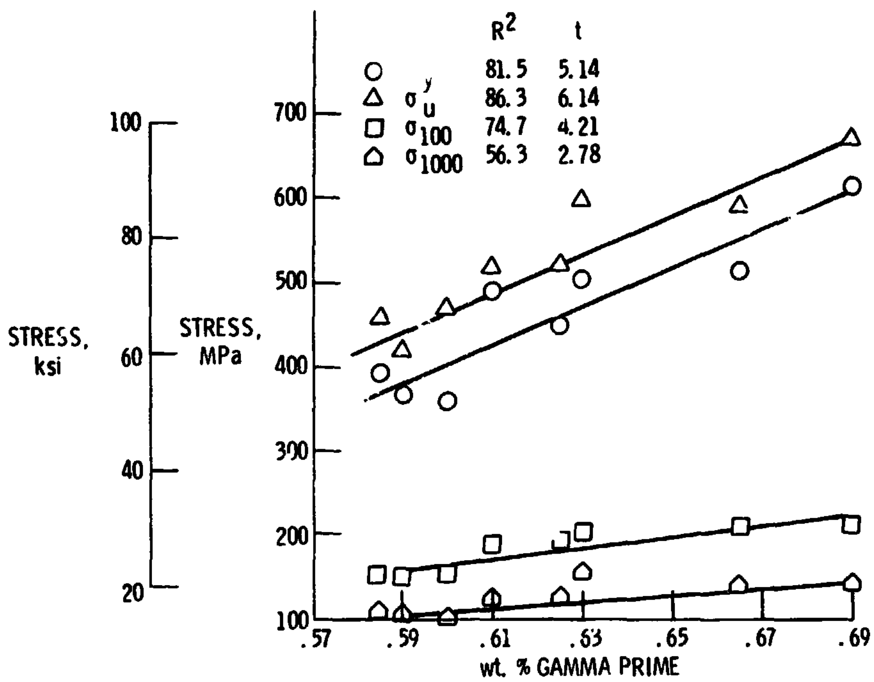
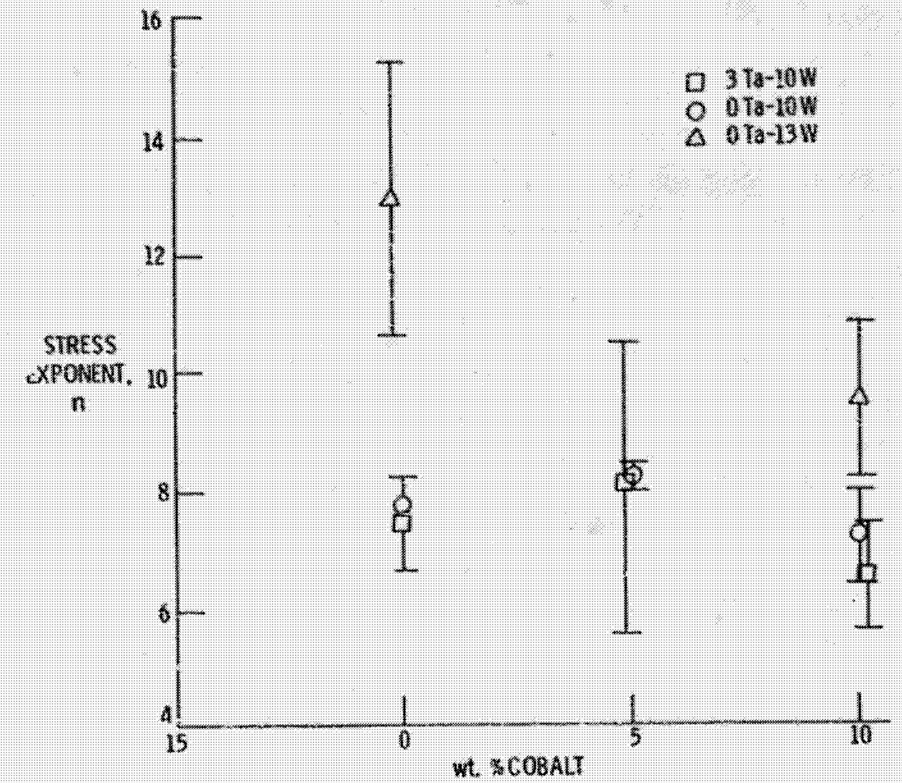
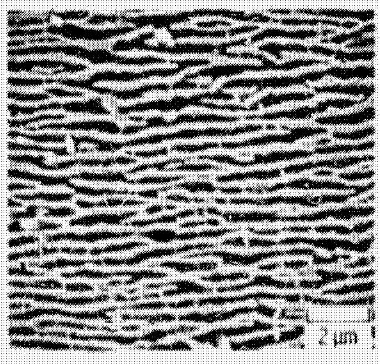
1000° C STRESS CAPABILITY OF [100] ORIENTED SINGLE CRYSTALS AS A FUNCTION OF wt. % GAMMA PRIME



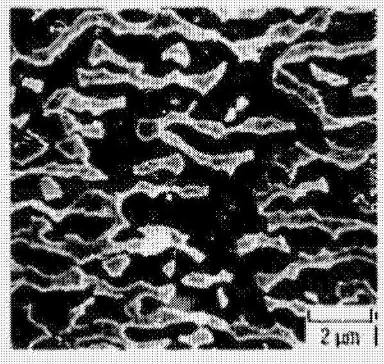
1000° C STRESS EXPONENTS OF [100] ORIENTED SINGLE CRYSTALS AS A FUNCTION OF COMPOSITION



ORIENTED COARSENING OF GAMMA PRIME DURING CREEP DEFORMATION AT 1000° C AND 148 MPa



TEST INTERRUPTED AT t = 20
ε = 0.005, t/t$_f$ = 0.025



FAILED SPECIMEN,
ε_f = 0.15, t_f = 790 hr

CS-82-2242

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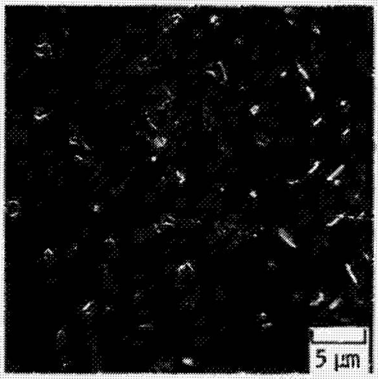
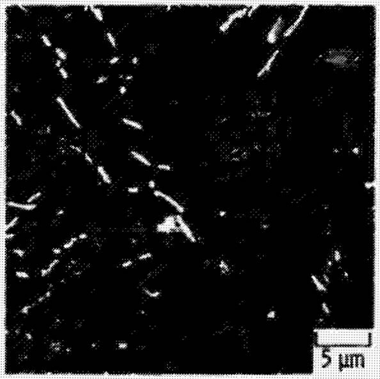
ALPHA-TUNGSTEN AND MU PHASES FOUND IN ALLOY B (0 Co-3 Ta-10 W)



SOLUTION TREATED

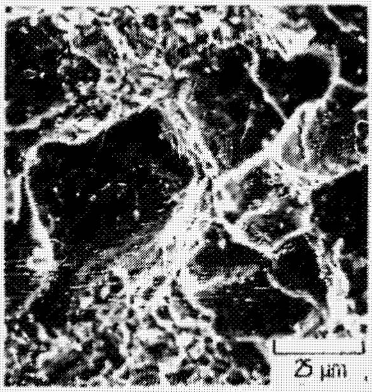


FAILED RUPTURE SPECIMEN,
1000° C, 207 MPa, t_f = 790 hr

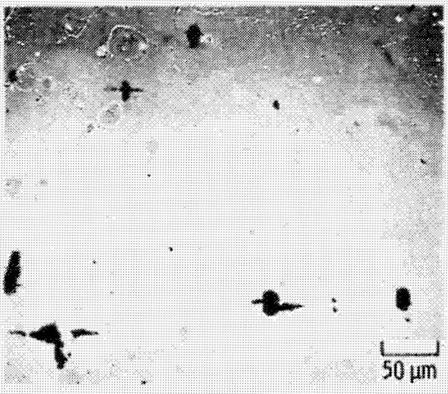
CS-82-2244

MICROSTRUCTURE OF FAILED CREEP-RUPTURE SPECIMEN, ALLOY B (0 Co-3 Ta-10 W)

1000° C, 207 MPa (30 ksi), t_f = 110 hr



FRACTURE SURFACE



LONGITUDINAL SECTION

CS-82-2243

SUMMARY

MICROSTRUCTURE OF SINGLE CRYSTAL ALLOYS

1. DECREASE Co: INCREASE IN γ' SOLVUS
   : INCREASE IN wt. % γ' (3 Ta-10 W)
   : INCREASE IN TCP PHASE FORMATION
2. SUBSTITUTE Ni FOR Ta: STRONG DECREASE IN γ' SOLVUS
   : STRONG DECREASE IN wt. % γ'
3. SUBSTITUTE W FOR Ta: SMALL DECREASE IN γ' SOLVUS
   : SMALL DECREASE IN wt. % γ'

MECHANICAL PROPERTIES OF SINGLE CRYSTAL ALLOYS

1. DECREASE Co: INCREASE IN CREEP RESISTANCE AND TENSILE STRENGTH FOR THE HIGH (Ta + W) LEVELS
   : VERY SMALL EFFECT FOR THE LOW (Ta + W) LEVELS
2. ALL ALLOYS HAD TENSILE ELONGATIONS GREATER THAN 18 %. LOWER STRENGTH ALLOYS HAD HIGHER DUCTILITY
3. TUNGSTEN IS MORE EFFECTIVE THAN Ta FOR CREEP RESISTANCE
4. 1000° C STRESS CAPABILITY IS STRONGLY CORRELATED WITH wt. % γ'
5. CASTING POROSITY APPEARS TO BE A MORE SERIOUS DEFECT THAN THE PRESENCE OF TCP PHASES

N83 11294 D/12

STRUCTURE-PROPERTY EFFECTS OF TANTALUM ADDITIONS TO NICKEL-BASE SUPERALLOYS

R. W. Heckel, B. J. Pletka, and D. A. Koss
Michighan Technological University
Houghton, Michigan

and

M. R. Jackson
General Electric Company
Schenectady, New York

The principal thrusts of this research effort are the characterization of the effect of Ta on the structure of Ni-base superalloys, the determination of the effects of Ta (structure) variations on the mechani al, thermal, and oxidation behavior, and the identification of alloying elements which have potential as substitutes for Ta. Primary attention is being directed toward Mar M247-type alloys; nominal and analyzed compositions of ten alloys currently under study are given (1-2).

X-ray and composition analysis are being used to determine the partitioning of alloying elements between γ, γ', and MC (cubic) as a function of Ta content. Preliminary data are given (3-8). These studies will continue on the remainder of the alloys as well as on additional compositions.

The diffusional interactions of the Mar M247-type alloys with as-cast $\beta+\gamma$ alloys are being studied to determine the effects of Ta on alloy/coating degradation. Preliminary data are given (9-10). Preliminary high-temperature oxidation data for the alloys are also given (11).

Forthcoming new research efforts will include high-temperature creep and tensile testing, plasma-spray coating studies, and cyclic oxidation testing. The concurrent (continuing) structure studies will provide the basis for structure/property correlations.

NOMINAL COMPOSITIONS

	A	B	C	D	E	F	G	H	I	J
	Conv	Conv	DS	DS	XL	XL	XL	XL	XL	XL
Ta	–	3	–	3	–	3	4.5	–	1.5	3
C	0.1	0.1	0.1	0.1	0.1	0.1	0.1	–	–	–
Zr	.05	.05	.05	.05						
B	.01	.01	.01	.01						
Hf	1.2	1.2	1.2	1.2						

For all alloys:
Cr 8.0, W 9.6, Co 9.5, Mo 0.5, Al 5.2, Ti 0.6, Ni Bal

ALLOY ANALYSES

	A	B	C	D	E	F	G	H	I	J
	Conv	Conv	DS	DS	XL	XL	XL	XL	XL	XL
Ta			–	2.9	–	2.8	4.3	–	1.6	3.1
C			0.10	0.10	0.11	0.10	0.11	0.01	0.01	0.01
Zr			0.06	0.06	0.01	0.01	0.01	0.01	0.01	0.01
B			0.01	0.01	–	–	–	–	–	–
Hf			1.2	1.2	–	–	–	–	–	–
Cr			8.3	8.4	7.9	8.0	8.0	7.9	8.0	8.1
W			9.4	9.6	9.5	9.8	9.8	9.6	9.6	9.7
Co			9.5	9.6	9.1	9.6	9.7	9.2	9.4	9.6
Mo			0.5	0.6	0.5	0.6	0.6	0.5	0.5	0.6
Al			4.9	5.4	4.9	5.4	5.6	4.9	5.2	5.5
Ti			0.8	0.8	0.5	0.5	0.6	0.5	0.6	0.6

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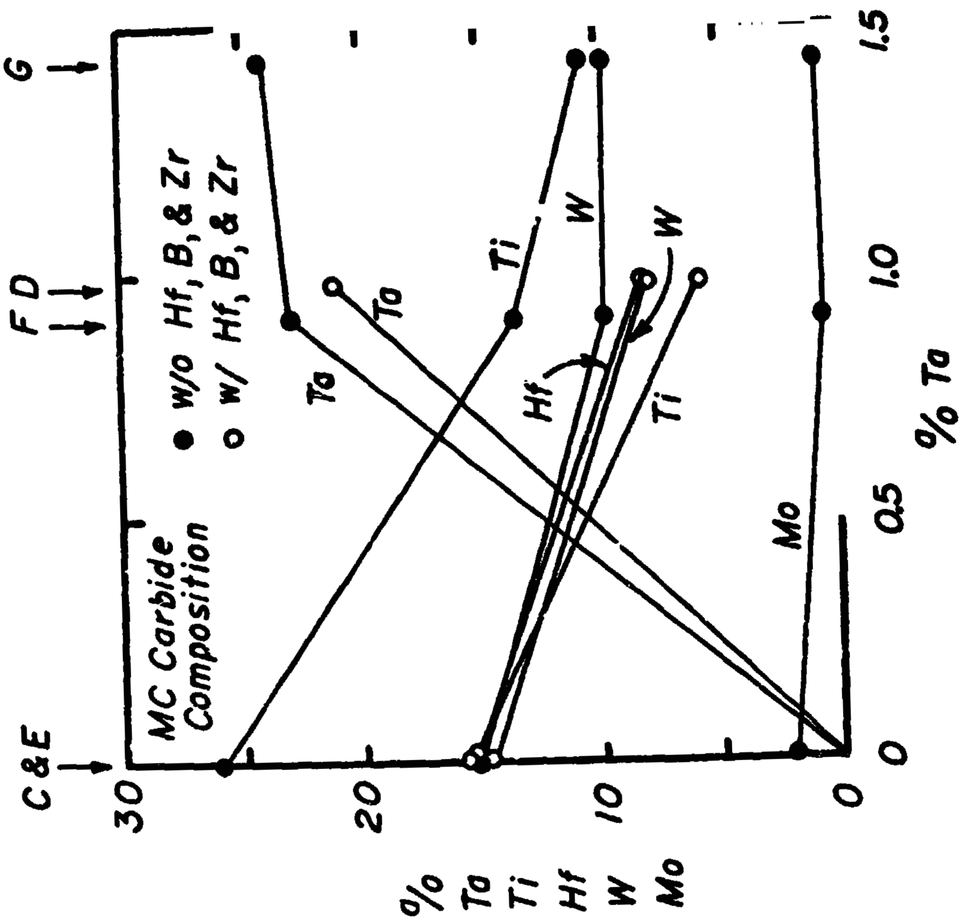
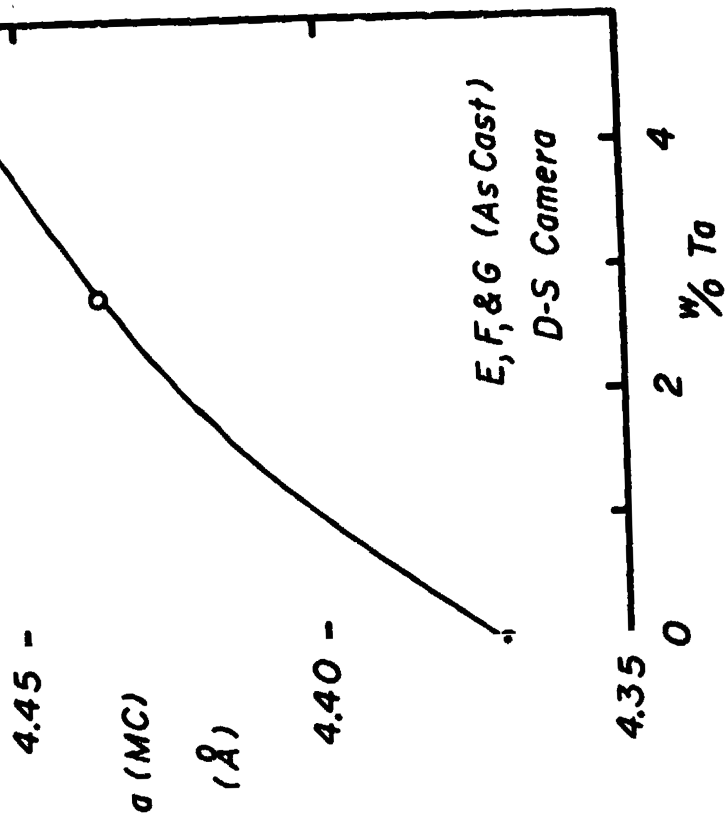
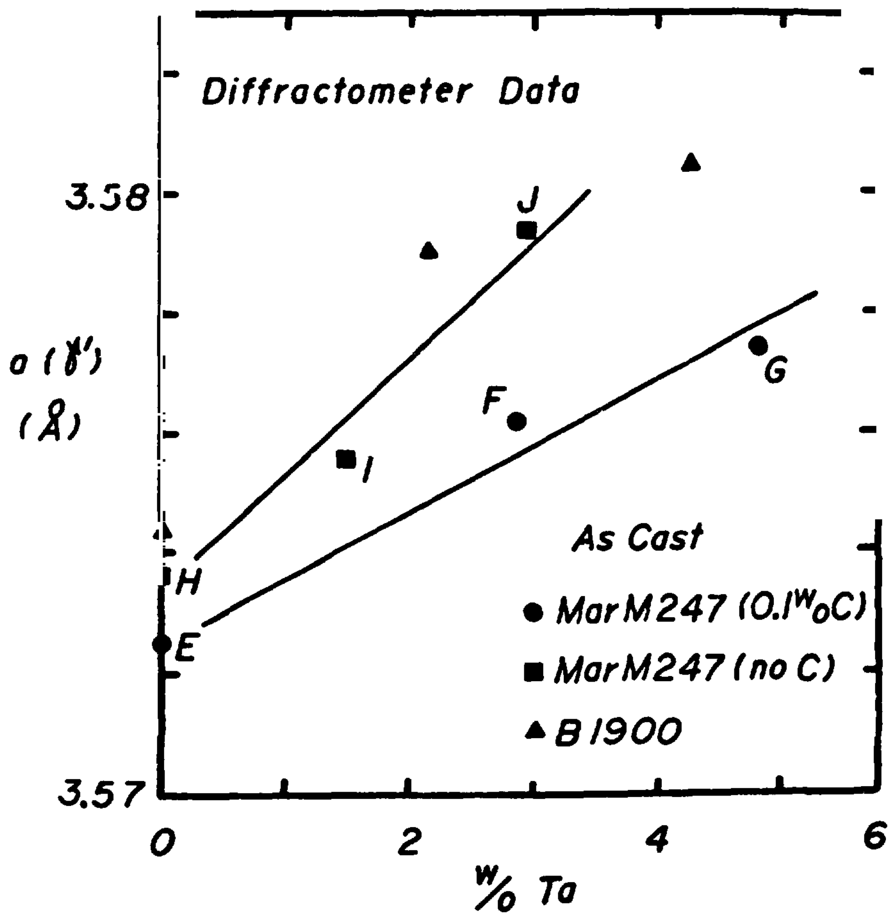
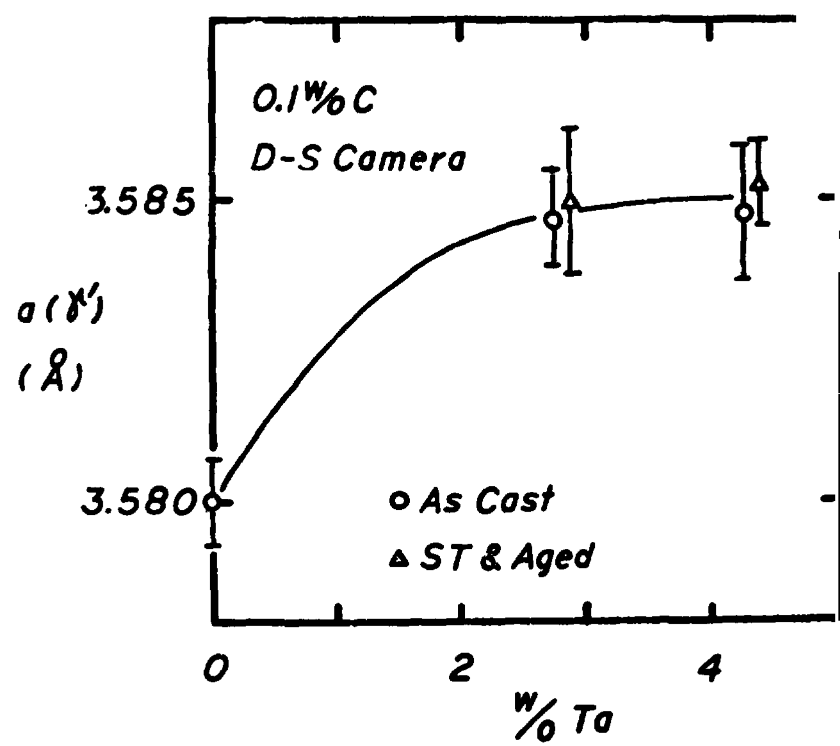






Diffractometer Data
3.58
3.57
$a(\gamma')$
$(\mathring{A})$
J
I
F
G
H
E
As Cast
● Mar M247 (0.1 w/o C)
■ Mar M247 (no C)
▲ B 1900
0
2
4
6
w/o Ta



0.1 w/o C
D-S Camera
3.585
3.580
$a(\gamma')$
$(\mathring{A})$
○ As Cast
△ ST & Aged
0
2
4
w/o Ta

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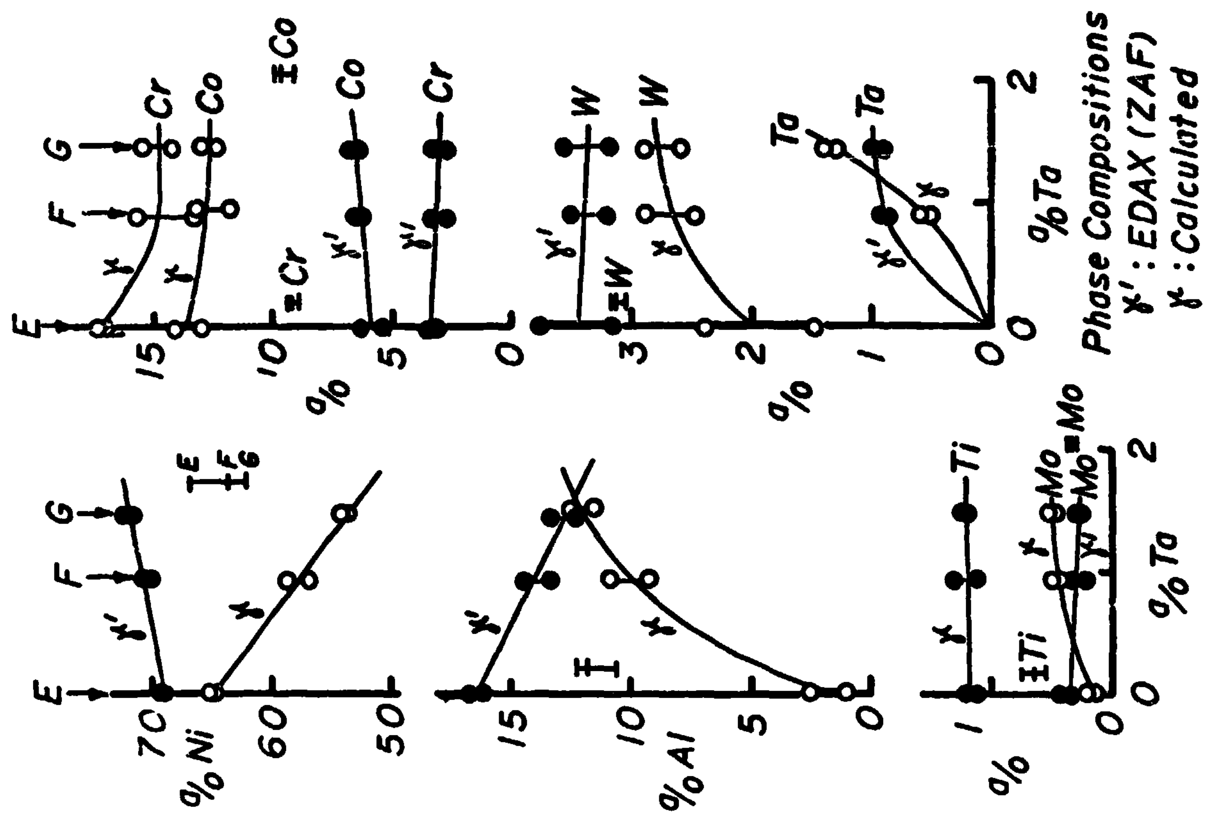
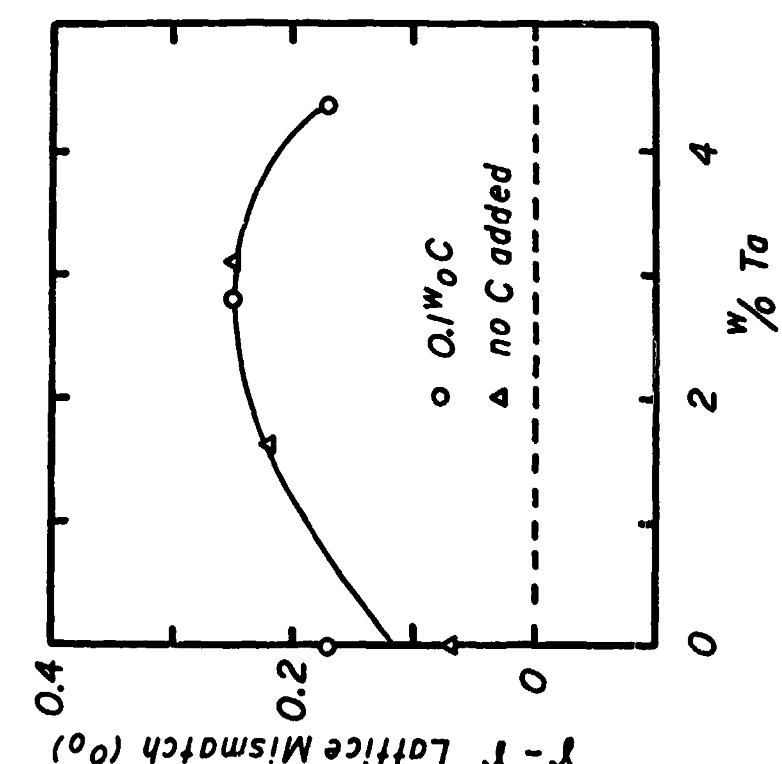
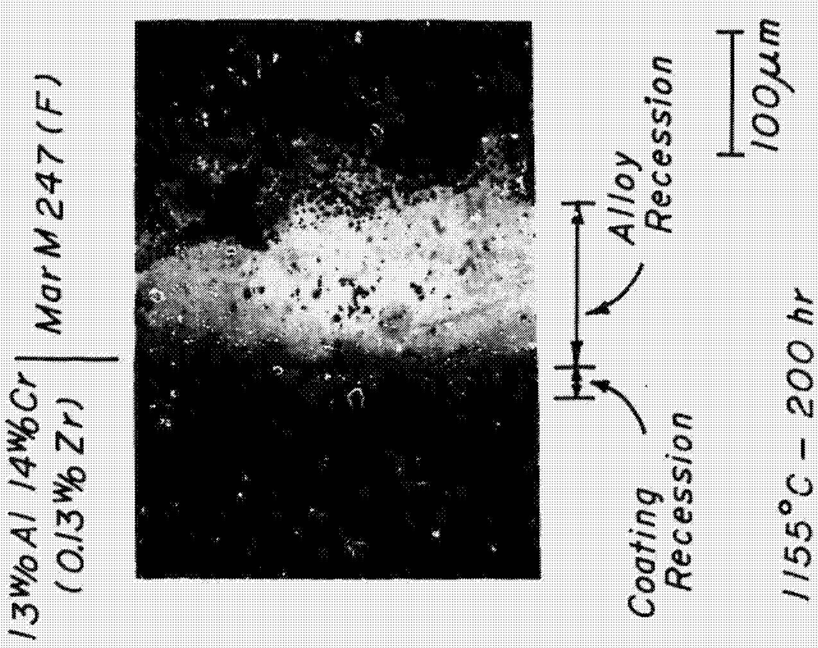
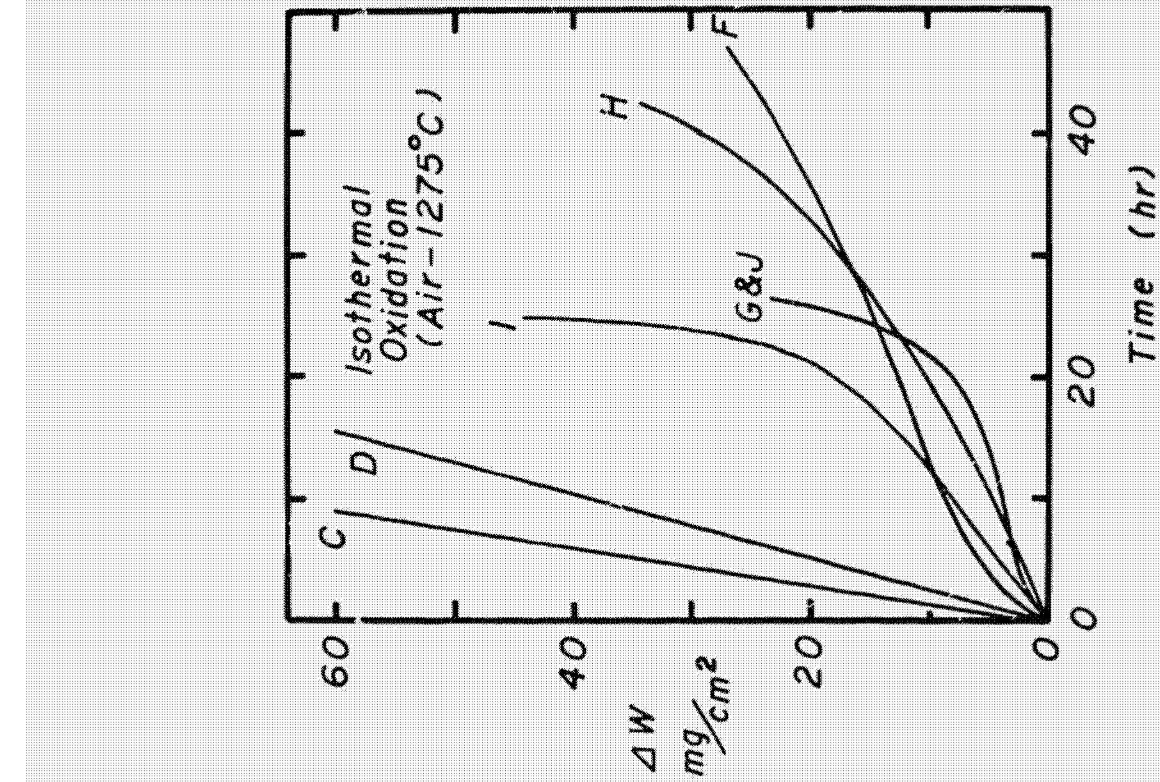





13 w/o Al 14 w/o Cr
(0.13 w/o Zr)
Mar M 247 (F)
Alloy
Recession
Coating
Recession
100 μm
1155°C – 200 hr


Isothermal
Oxidation
(Air–1275°C)
ΔW
mg/cm²
60
40
20
0
C
D
I
G&J
H
F
0
20
40
Time (hr)

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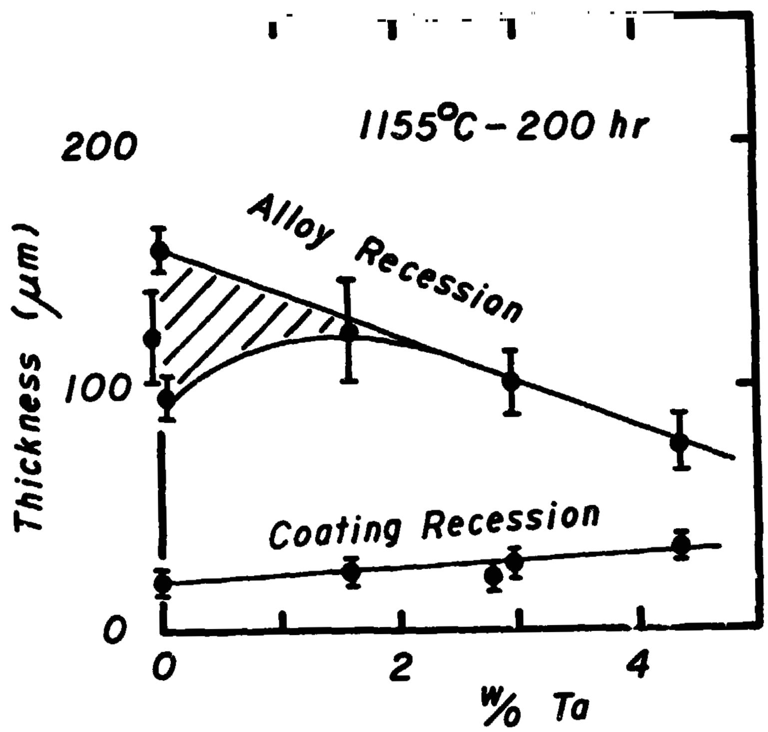


MECHANICAL PROPERTIES OF LOW TANTALUM ALLOYS

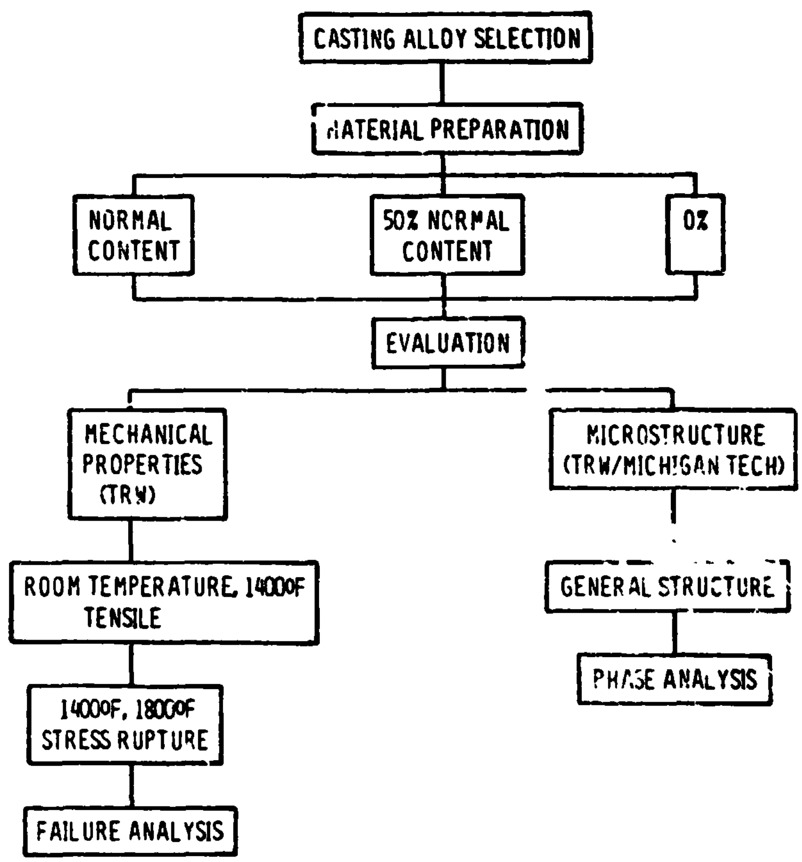
C. S. Kortovich
TRW, Inc.
Cleveland, Ohio

A study was performed on the mechanical property behavior of equiaxed cast B-1900 + Hf alloy as a function of tantalum content. Tensile and stress rupture characterization was conducted on cast to size test bars containing tantalum at the 4.3% (standard level), 2.2% and 0% levels.

Casting parameters were selected to duplicate conditions used to prepare test specimens for master metal heat qualification. The mechanical property results as well as results of microstructural/phase analysis of failed test bars will be presented.

MECHANICAL PROPERTIES OF LOW TANTALUM ALLOYS

OBJECTIVE: DEVELOP IMPROVED UNDERSTANDING OF INTERACTIONS OF TANTALUM IN CAST B1900 PLUS HAFNIUM SUPERALLOY



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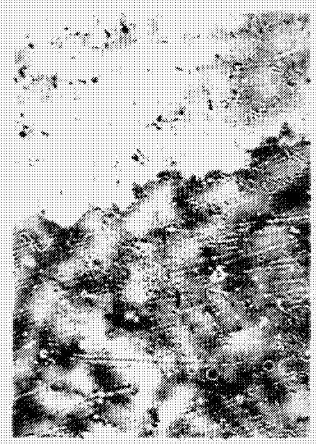
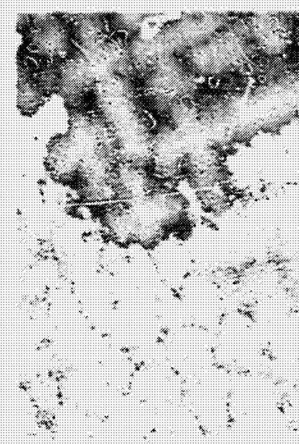
GENERAL STRUCTURE AT 100X MAGNIFICATION

4.3%Ta



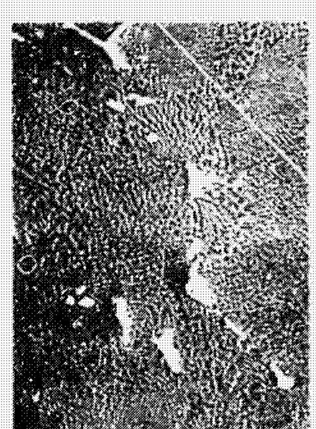
2.2%Ta



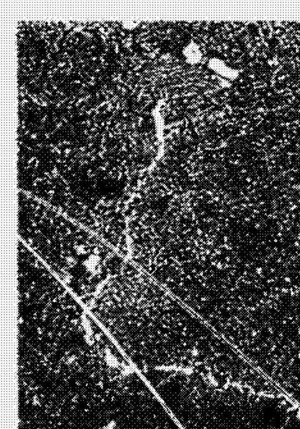
0%Ta

GENERAL STRUCTURE AT 1000X MAGNIFICATION

4.3%Ta



2.2%Ta



0%Ta

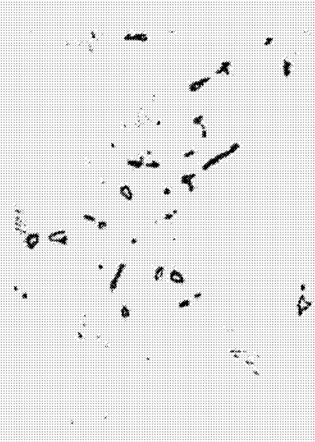
CARBIDE DISTRIBUTION AT 100X MAGNIFICATION

4.3%Ta

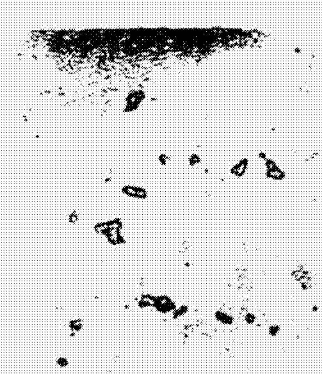
2.2%Ta

0%Ta

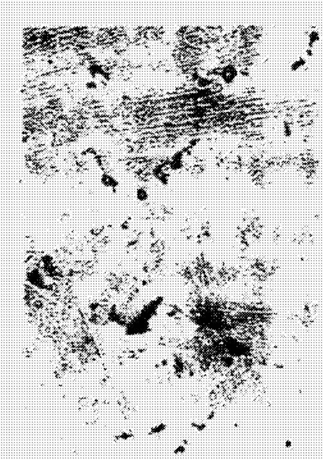
CARBIDE DISTRIBUTION AT 500X MAGNIFICATION



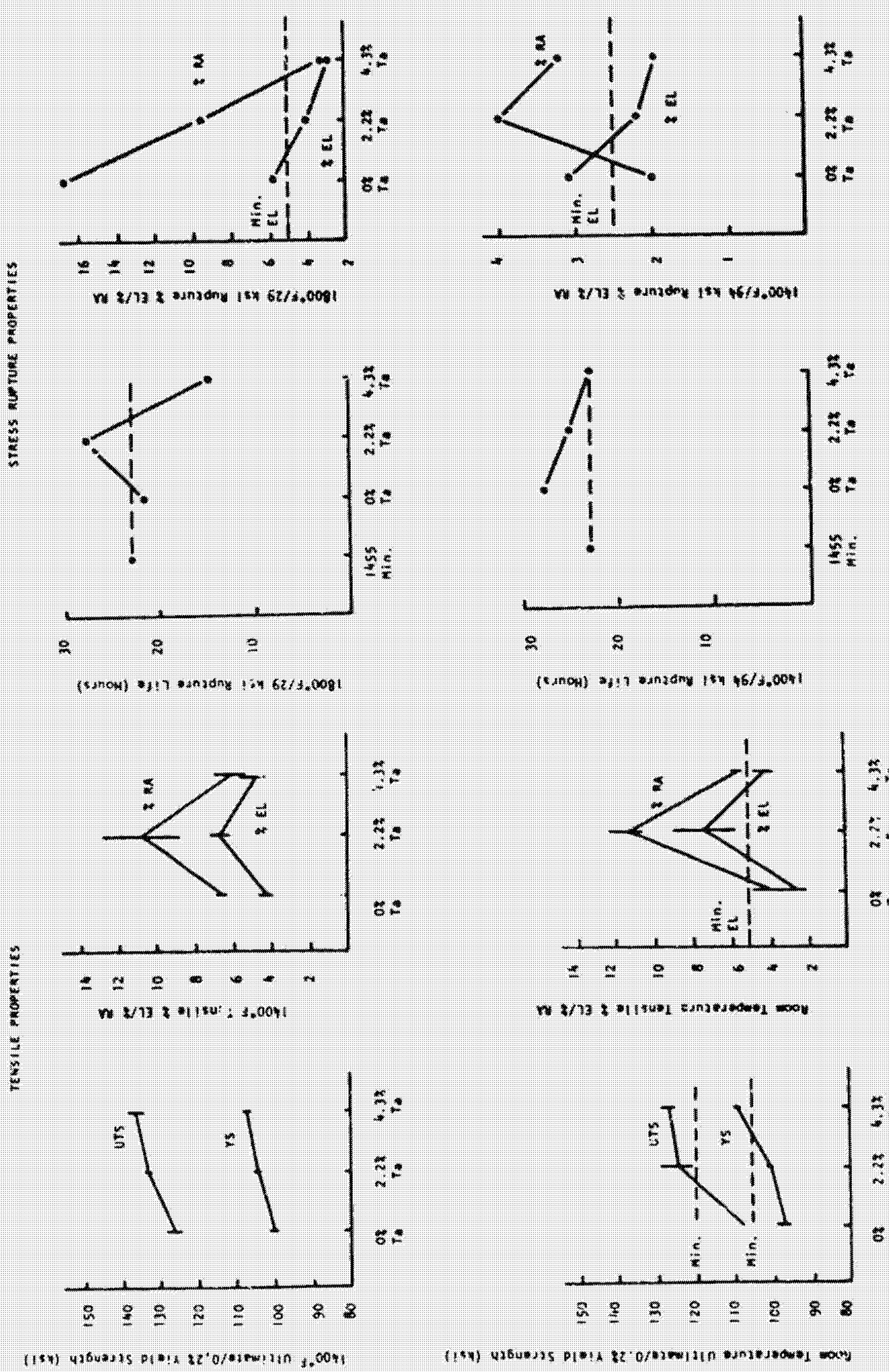
4.3%Ta



2.2%Ta



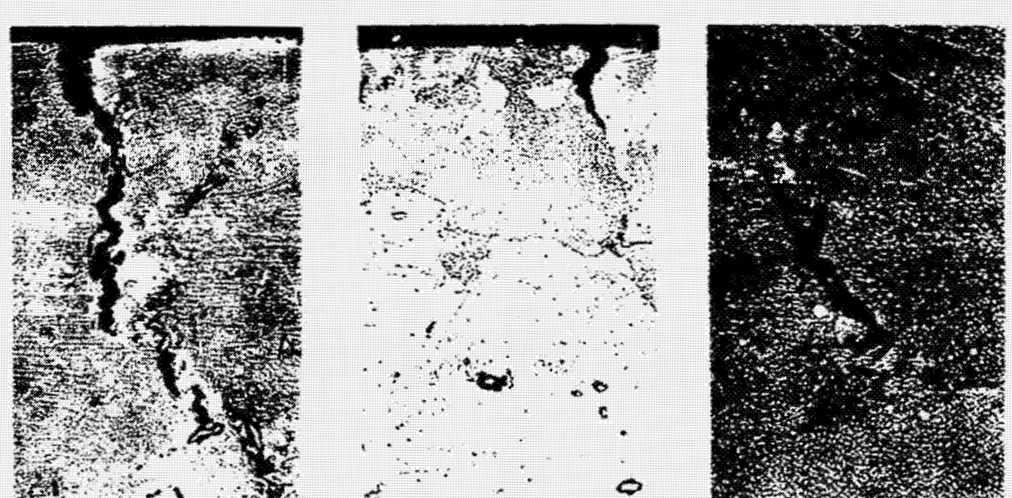
0%Ta


STRESS RUPTURE PROPERTIES
TENSILE PROPERTIES
1800°F/29 ksi Rupture % EL/% RA
1400°F/94 ksi Rupture % EL/% RA
1800°F/29 ksi Rupture Life (Hours)
1400°F/94 ksi Rupture Life (Hours)
1400°F Tensile % EL/% RA
Room Temperature Tensile % EL/% RA
1400°F Ultimate/0.2% Yield Strength (ksi)
Room Temperature Ultimate/0.2% Yield Strength (ksi)
% RA
% EL
Min. EL
UTS
YS
Min.
1455 Min.
0% Ta
2.2% Ta
4.3% Ta

1400°F TENSILE FAILURE 500X

4.3%Ta 2.2%Ta 0%Ta

1400°F STRESS RUPTURE FAILURE 500X



4.3%Ta 2.2%Ta 0%Ta

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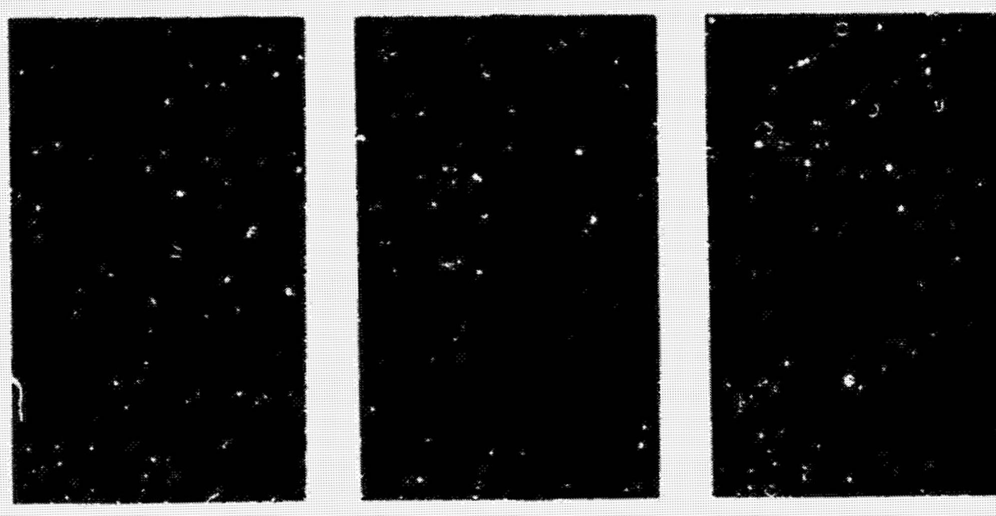
1800°F STRESS RUPTURE FAILURE 500X



4.3%Ta 2.2%Ta 0%Ta

SUMMARY

- Microstructural Characteristics
  - Reduced Ta Changes Eutectic Colony Formation From Spherulitic to Lamellar Morphology
  - Reduced Ta Reduces Amount of Carbide Formation
  - Reduced Ta Eliminates Script Carbide Formation
- Mechanical Property Characteristics
  - Reduced Ta Results in Loss of Ultimate Strength at Yield Strength at Room Temperature and 1400°F
  - Tensile Ductility is Optimum and 50% Normal Ta Content
  - Reduced Ta Results in Improved 1400°F and 1800°F Rupture Life and Ductility
- Failure Characteristics
  - Fracture Path is Intergranular/Interdendritic
  - With Reduced Ta, Fracture Path Follows Edge of Eutectic Colonies but Goes Thru Blocky Carbides
  - With Normal Ta, Fracture Path Goes Thru Eutectic Colonies
  - Script Carbides not Associated With Fracture Path

N83 11296 -9/4

EFFECT OF REDUCTION OF STRATEGIC COLUMBIUM ADDITIONS IN INCONEL 718 ALLOY ON THE STRUCTURE AND PROPERTIES

Karl Ziegler and John F. Wallace
Department of Metallurgy and Materials Science
Case Western Reserve University
Cleveland, Ohio 44106

This investigation is designed to determine whether 3%W, combinations of 3%W and 0.9%V, raising Mo to 5.8% from its normal value of 3% or increasing B to 0.04% can be employed to reduce Cb to 3% or 1% from its normal 5.2% in Inconel 718 alloy. A series of twelve alloy combinations of hot rolled 0.5 inch thick sections with various combinations of Cb, W, V, Mo and B within these limits and containing the usual other elements have been prepared by Special Metals. These have been solution heat treated at temperatures of 1700, 1800, 1900 and 2000°F and aged for various times at 1200, 1300, 1400, 1500 and 1600°F. This amounts to a total of 40 treatments of 12 alloys or 480 tests. The structure is being examined after these treatments and those treatments and alloy-treatments that show promise will be tested to determine their tensile and stress rupture properties at 1000, 1100 and 1200°F.

OBJECTIVE

To determine how much columbium can be removed from Inconel Alloy 718 without degrading its high temperature properties. The elements that are being substituted are: vanadium and tungsten together and separately; increasing the molybdenum level from 3.0% to 5.8% and increasing the boron to 0.04%.

The following solution treatment temperatures were used (in °F)

2000
1900
1800
1700

For each solution temperature the following aging temperatures were used:

TEMPERATURE °F	TIME (HRS)
1600	5
	10
1500	10
	25
1400	25
	50
	100
1300	50
	100
1200	100

These treatments are nearly completed.

Selection for mechanical testing will depend on the analysis of the structures.

There is a total of: 40 treatments x 12 compositions = 480 structures.

ALLOY	Cb+Te	Mo	V	W	B
1	5.32	3.10	-	-	-
2	5.30	3.10	-	-	0.04
3	3.20	3.10	-	-	-
4	3.10	5.80	-	-	-
5	3.00	2.99	-	3.0	-
6	3.00	3.00	0.9	3.0	-
7	3.10	3.10	-	-	0.04
8	3.10	5.80	-	-	0.04
9	1.10	3.10	-	-	-
10	1.10	5.80	-	-	-
11	1.10	3.00	-	3.0	-
12	1.10	3.00	0.9	3.0	-

Elements Common to All Alloys:

Aluminum	0.4 - 0.8	Silicon	0.35 Max.
Titanium	0.65 - 1.15	Phosphorous	0.15 Max.
Chromium	17.0 - 21.0	Sulfur	0.15 Max.
Carbon	0.1 Max.	Iron	18.0 - 20.0
Manganese	0 35 Max.	Nickel + Cobalt	Balance

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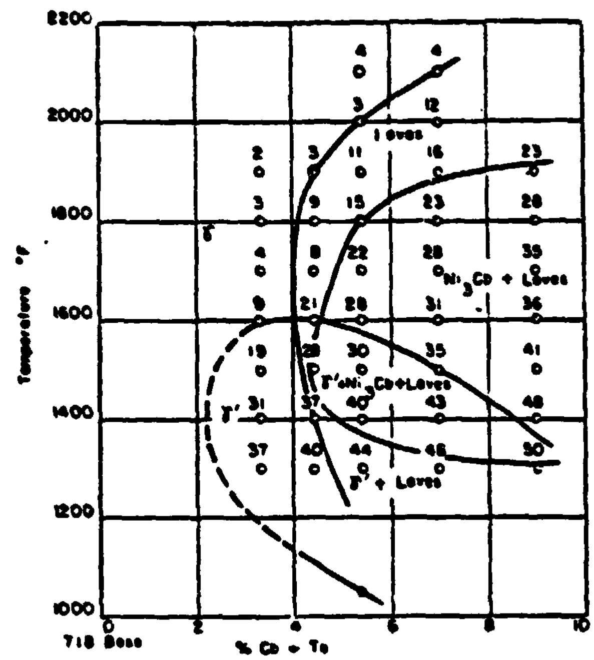

FIGURE 1 - Composite Phase Diagram Shows the occurrence of Various Phases After Heating for 100 Hr at the Temperatures Indicated. (Rockwell "C" hardness also is shown for the heat treated specimens.

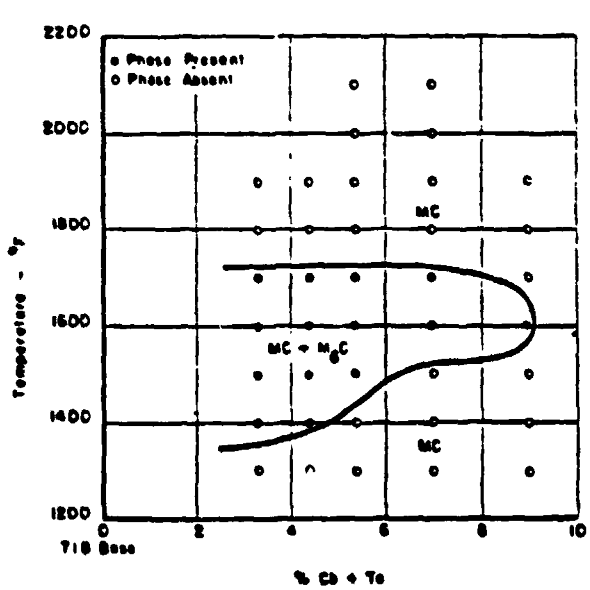

FIGURE 2 - Phase Diagram for M_6C Developed by Heating Specimens of Various Percentages of Cb Plus Ta for 100 Hr at the Temperatures Indicated.

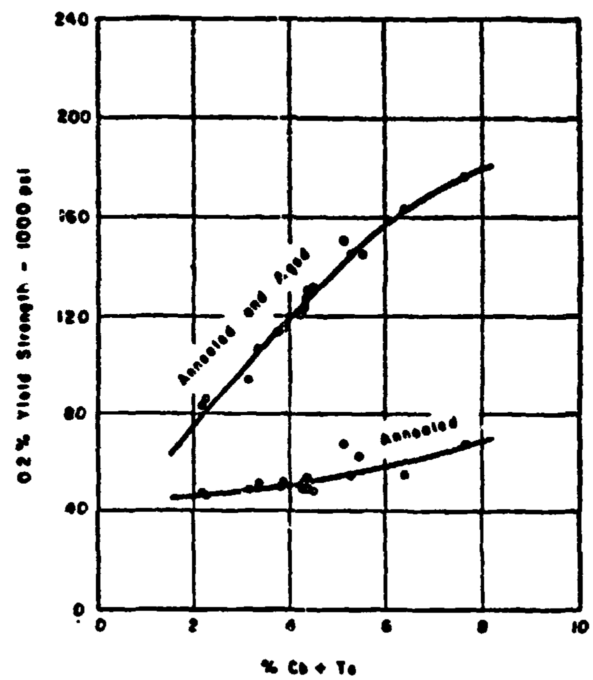

FIGURE 3 - Effect of Per Cent Cb Plus Ta on 0.2 offset percent Yield Strength: Material Was Annealed at 1900 F/1 Hr Water Quenched, Aged at 1250 to 1350 F/16 Hr, Air Cooled.

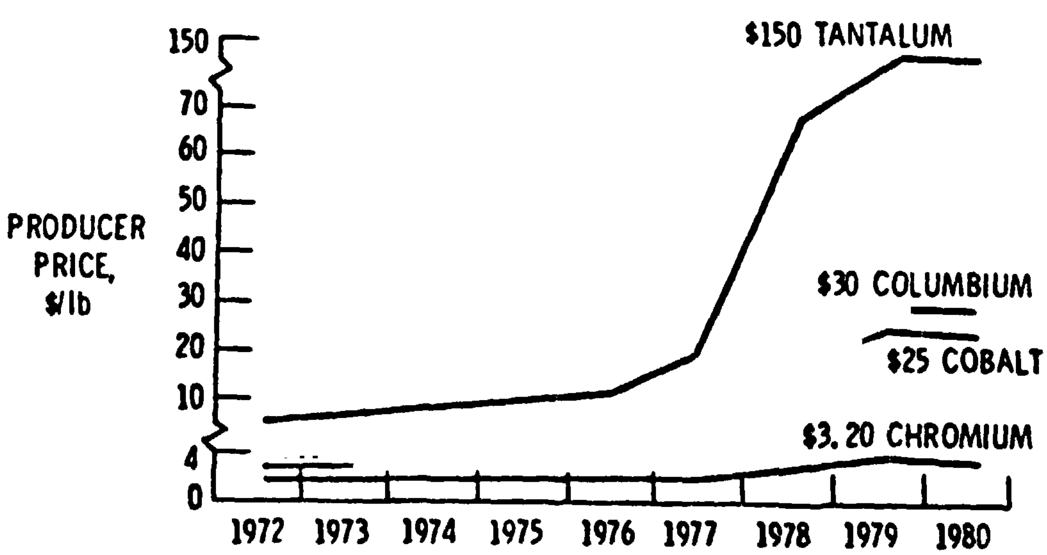

FIGURE 4 - Cost increase of selected strategic metals over the past nine years.

FIGURE 5: ALLOY NO. 1 (STANDARD INCONEL 718) SOLUTION 2000°F 2 HRS.; AGE 1600°F 10 HRS.; 300X (HCl: H_2SO_4):HNO_3 ETCH)

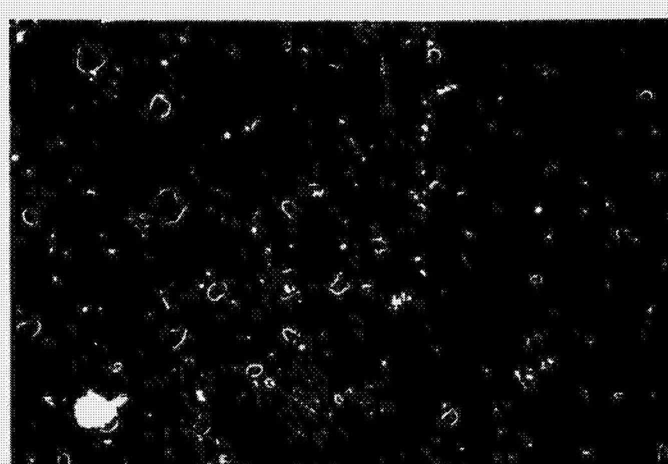

FIGURE 6: ALLOY NO. 1 (STANDARD INCONEL 718) SOLUTION 1900°F 2 HRS.; AGE 1400°F 100 HRS.; 300X (HCl: H_2SO_4: HNO_3 ETCH)

FIGURE 7: ALLOY NO. 1 (STANDARD INCONEL 718) SOLUTION 1700°F 2 HRS.; AGE 1200°F 100 HRS.; 300X (HCl: H_2SO_4: HNO_3 ETCH)

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FIGURE 8: ALLOY NO. 6 (3.0 Cb+Ta, 3.0 Mo, .9V, 3.0W) SOLUTION 2000°F 2 HRS.; AGE 1600°F 10 HRS.; 300X (HCl: H_2SO_4: HNO_3 ETCH)

FIGURE 9: ALLOY NO. 6 (3.0 Cb+Ta, 3.0 Mo, .9V, 3.0W) SOLUTION 1900°F 2 HRS.; AGE 1400°F 100 HRS.; 300X (HCl: H_2SO_4: HNO_3 ETCH)

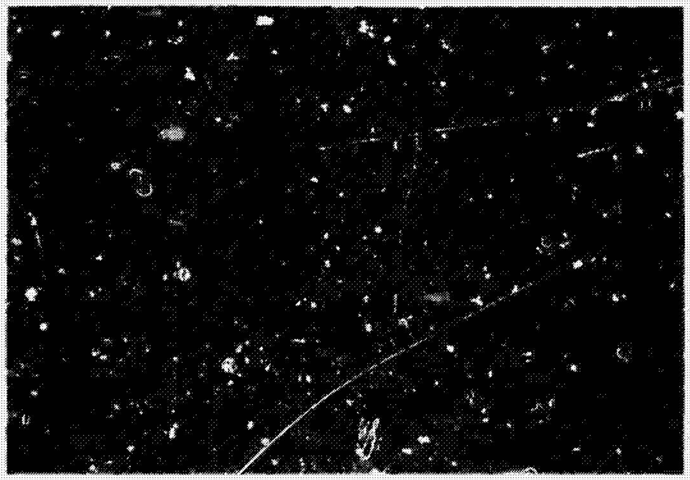

FIGURE 10: ALLOY NO. 6 (3.0 Cb+Ta, 3.0 Mo, .9V, 3.0W) SOLUTION 1700°F 2 HRS.; AGE 1200°F 100 HRS.; 300X (HCl: H_2SO_4: HNO_3 ETCH)

FIGURE 11: ALLOY NO. 12 (1.1 Cb+Ta, 3.0 Mo, .9V, 3.0W) SOLUTION 2000°F 2 HRS.; AGE 1600°F 10 HRS.; 300X (HCl: H_2SO_4: HNO_3 ETCH)

FIGURE 12: ALLOY NO. 12 (1.1. Cb+Ta, 3.0 Mo, .9V, 3.0W) SOLUTION 1900°F 2 HRS.; AGE 1400°F 100 HRS.; 300X (HCl: H_2SO_4: HNO_3 ETCH)

FIGURE 13: ALLOY NO. 12 (1.1 Cb+Ta, 3.0 Mo, .9V, 3.0W) SOLUTION 1700°F 2 HRS.: AGE 1200°F 100 HRS.; 300X (HCl: H_2SO_4: HNO_3 ETCH)

N83 11297 D/15

DUAL ALLOY INTERFACE STABILITY

Fredric H. Harf
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

The concept of powder metallurgy dual alloy fabrication is being applied to combinations of superalloys having a high iron, and low strategic metal content, with standard nickel-base superalloys, containing the strategic metals chromium, cobalt, and columbium. This program investigates the possibility of combining Alloy 901 (12 percent Cr, 36 percent Fe, 0 percent Co, and 0 percent Cb) with turbine disk alloys René 95 (13 percent Cr, 8 percent Co, and 4 percent Cb) or Low Carbon Astroloy (L.C.A.; 15 percent Cr, 17 percent Co, and 0 percent Cb). Preliminary results for combinations show that a strong interface with rapid diffusion is obtained between alloys and that the standard heat treatments for either alloy may be satisfactory.

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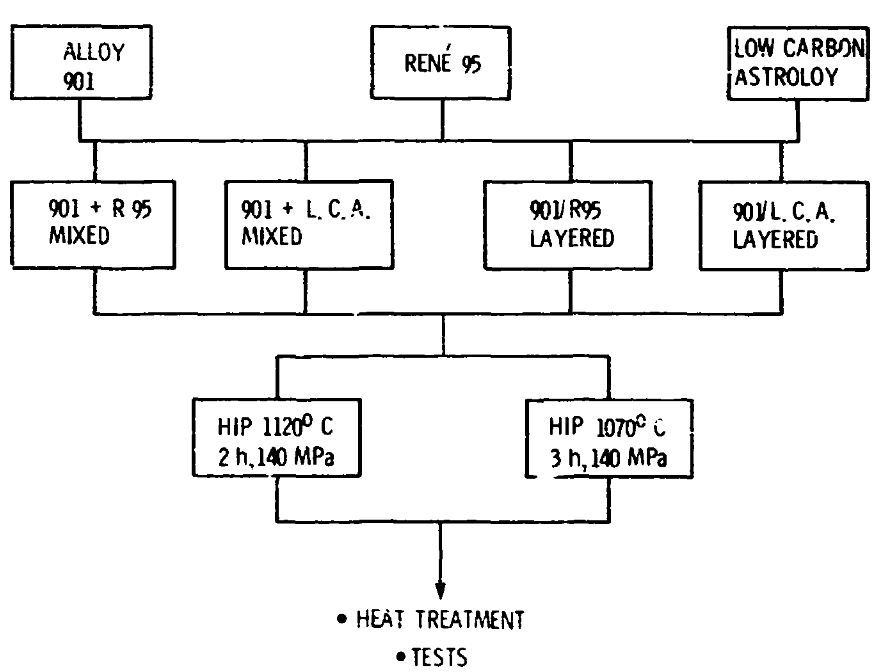
DUAL ALLOY INTERFACE STABILITY

- CONSERVE STRATEGIC MATERIALS
- DETERMINE THE COMPATIBILITY OF HIGH IRON CONTENT SUPERALLOYS WITH STANDARD NICKEL-BASE SUPERALLOYS IN DUAL ALLOY JOINTS PRODUCED FROM HOT ISOSTATICALLY PRESSED POWDERS
- EXTEND TECHNOLOGY OF DUAL ALLOY PROCESSING

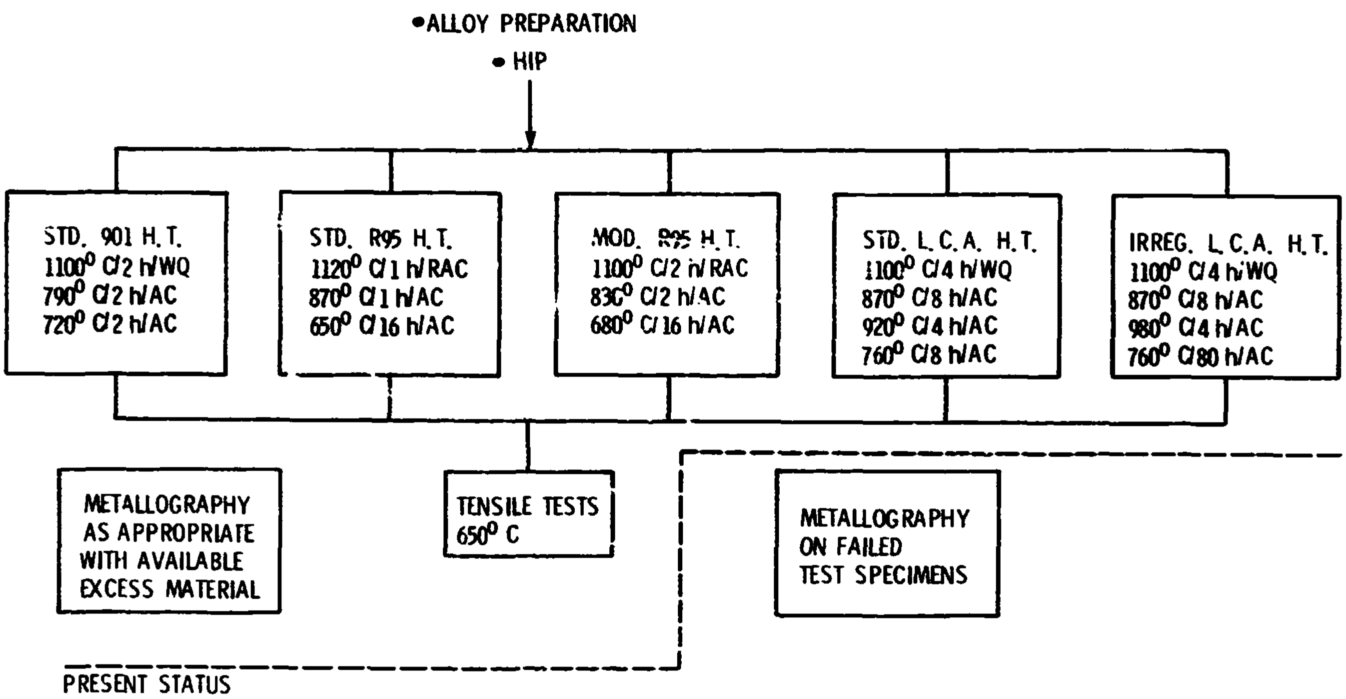
DUAL ALLOY INTERFACE STABILITY

BASIC HIP AND HEAT TREAT STUDY



DUAL ALLOY INTERFACE STABILITY

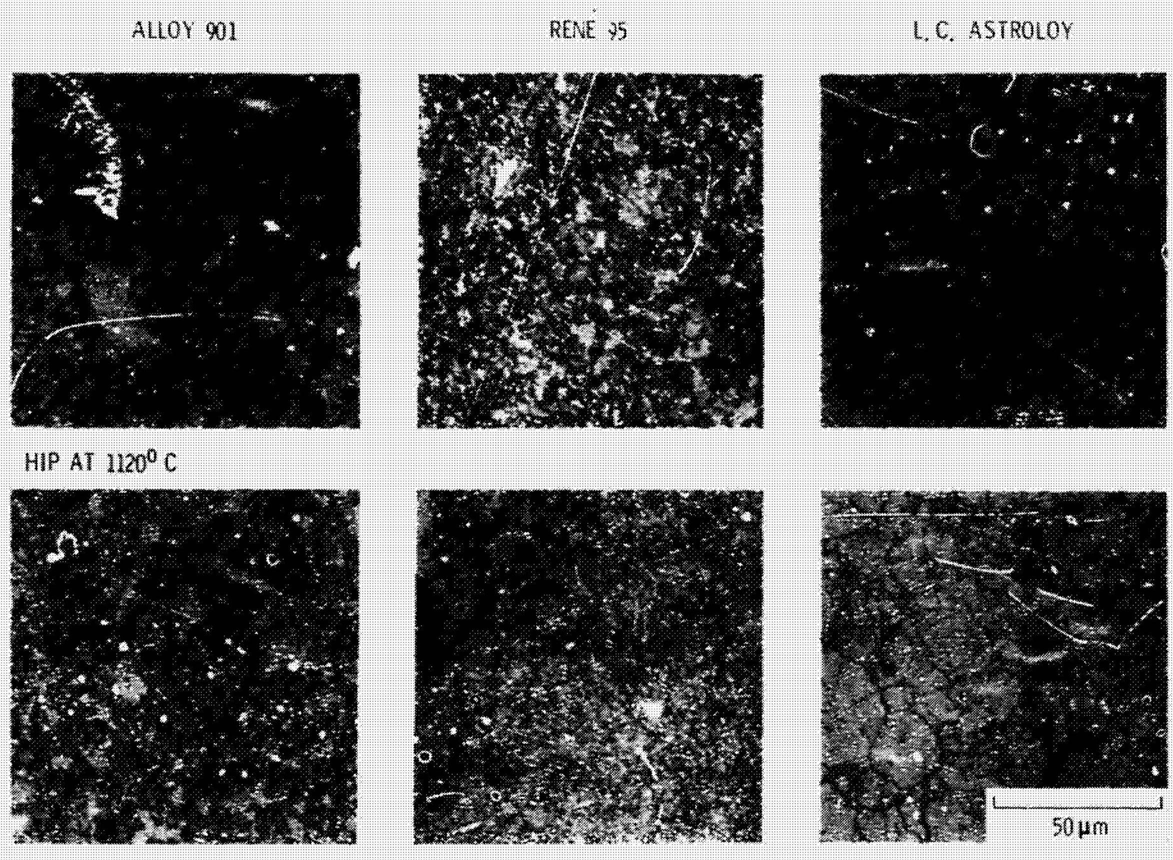
BASIC HIP AND HEAT TREATMENT STUDY



ALLOY COMPOSITIONS (ACTUAL)

	ALLOY 901	RENÉ 95	L.C. ASTROLOY
Fe	BAL	0.33	0.12
Ni	43.66	BAL	BAL
Cr	11.91	13.49	15.0
Mo	5.77	3.42	5.00
W		3.38	
Cb		3.70	
Co	0.06	7.90	17.09
Al	0.04	3.65	4.05
Ti	2.58	2.57	3.45
C	0.07	0.06	0.05
Zr		0.06	0.01
B	0.02	0.01	0.02

MICROSTRUCTURES OF AS HIP BASE ALLOYS

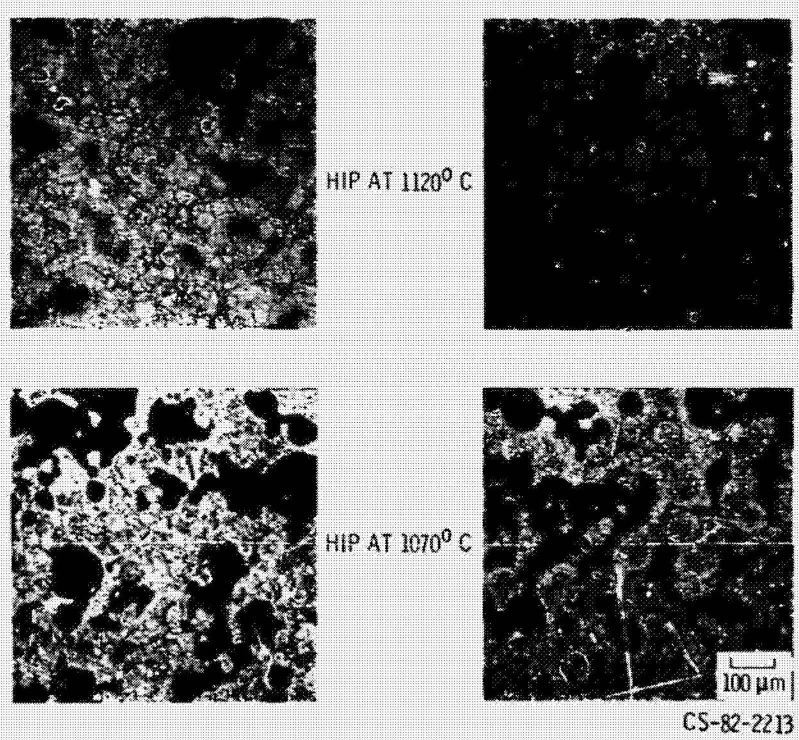


HIP at 1070° C

CS-82-2220

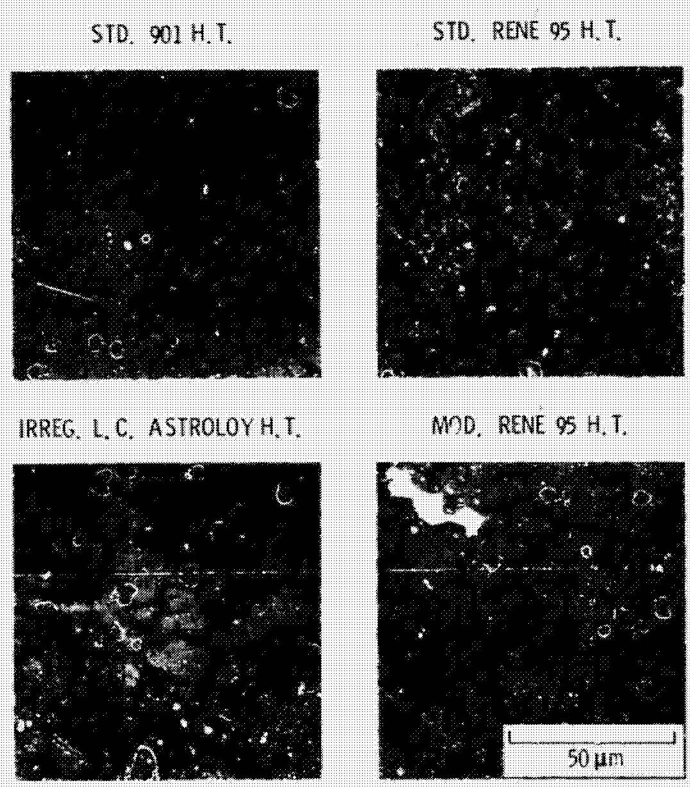
MICROSTRUCTURES OF AS HIP MIXED ALLOYS

ALLOY 901 + RENÉ 95 | ALLOY 901 + L.C. ASTROLOY



CS-82-2213

MICROSTRUCTURES OF HEAT TREATED ALLOY 901 (1120° C HIP)



CS-82-2216

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MICROSTRUCTURES OF HEAT TREATED ALLOY 901 EFFECT OF EXTENDED AGING (HIP AT 1070° C)

STD. L.C.A. HEAT TREATMENT
1100° C/4 h/WQ+
870° C/8 h/AC+
980° C/4 h/AC+
760° C/8 h/AC

IRREG. L.C.A. HEAT TREATMENT
1100° C/4 h/WQ+
870° C/8 h/AC+
980° C/4 h/AC+
760° C/80 h/AC

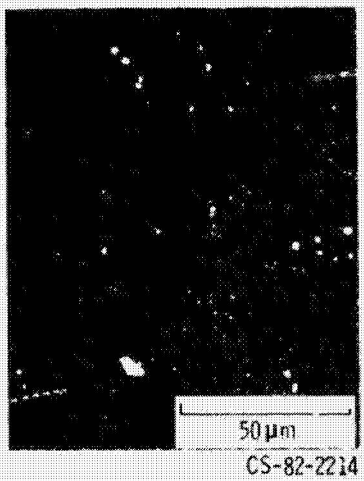
CS-82-2219

MICROSTRUCTURES OF HEAT TREATED RENÉ 95 (1120° C HIP)

STD. 901 H.T.

STD. RENÉ 95 H.T.

MOD. RENÉ 95 H.T.



CS-82-2214

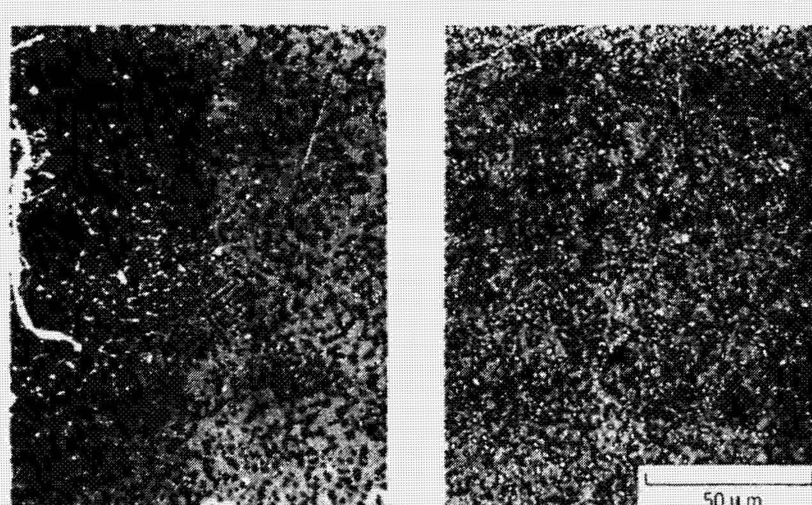
MICROSTRUCTURES OF HEAT TREATED MIXED ALLOY 901 +RENÉ 95 (1120° C HIP)

STD. 901 H.T. | STD. RENE 95 H.T. | MOD. RENÉ 95 H.T.

CS-82-2215

MICROSTRUCTURES OF HEAT TREATED LOW CARBON ASTROLOY (1120° C HIP)

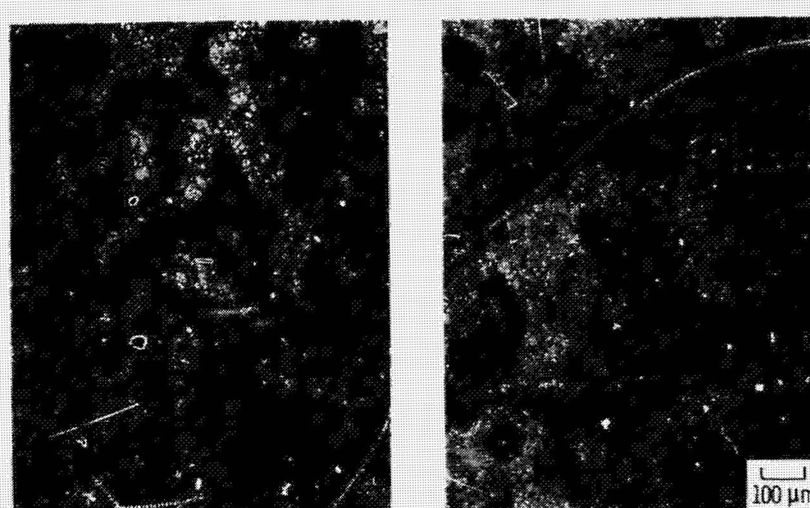
STD. 901 H.T. | IRREG. L.C. ASTROLOY H.T.



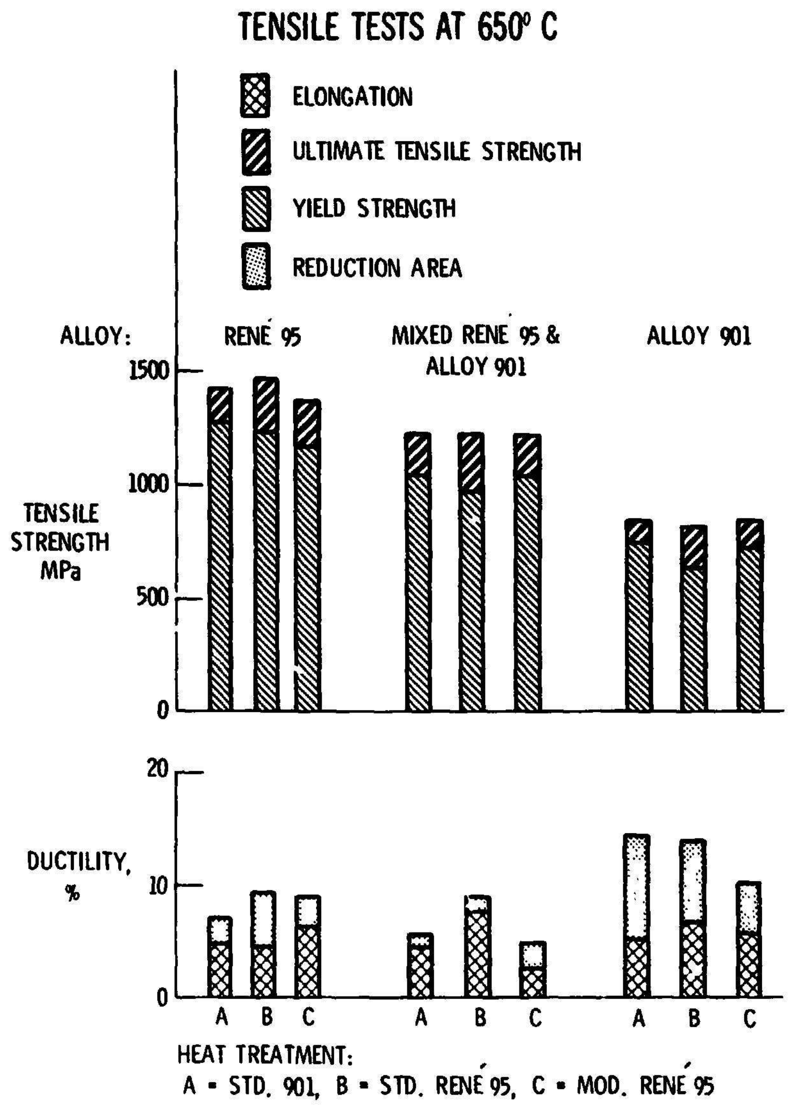
CS-82-2217

MICROSTRUCTURES OF HEAT TREATED MIXED ALLOY 901 + LOW CARBON ASTROLOY (1120° C HIP)

STD. 901 IRREG. H.C. ASTROLOY



CS-82-2218


TENSILE TESTS AT 650° C
ELONGATION
ULTIMATE TENSILE STRENGTH
YIELD STRENGTH
REDUCTION AREA
ALLOY:
RENÉ 95
MIXED RENÉ 95 & ALLOY 901
ALLOY 901
1500
1000
500
0
TENSILE STRENGTH MPa
20
10
0
DUCTILITY, %
A B C
A B C
A B C
HEAT TREATMENT:
A = STD. 901, B = STD. RENÉ 95, C = MOD. RENÉ 95

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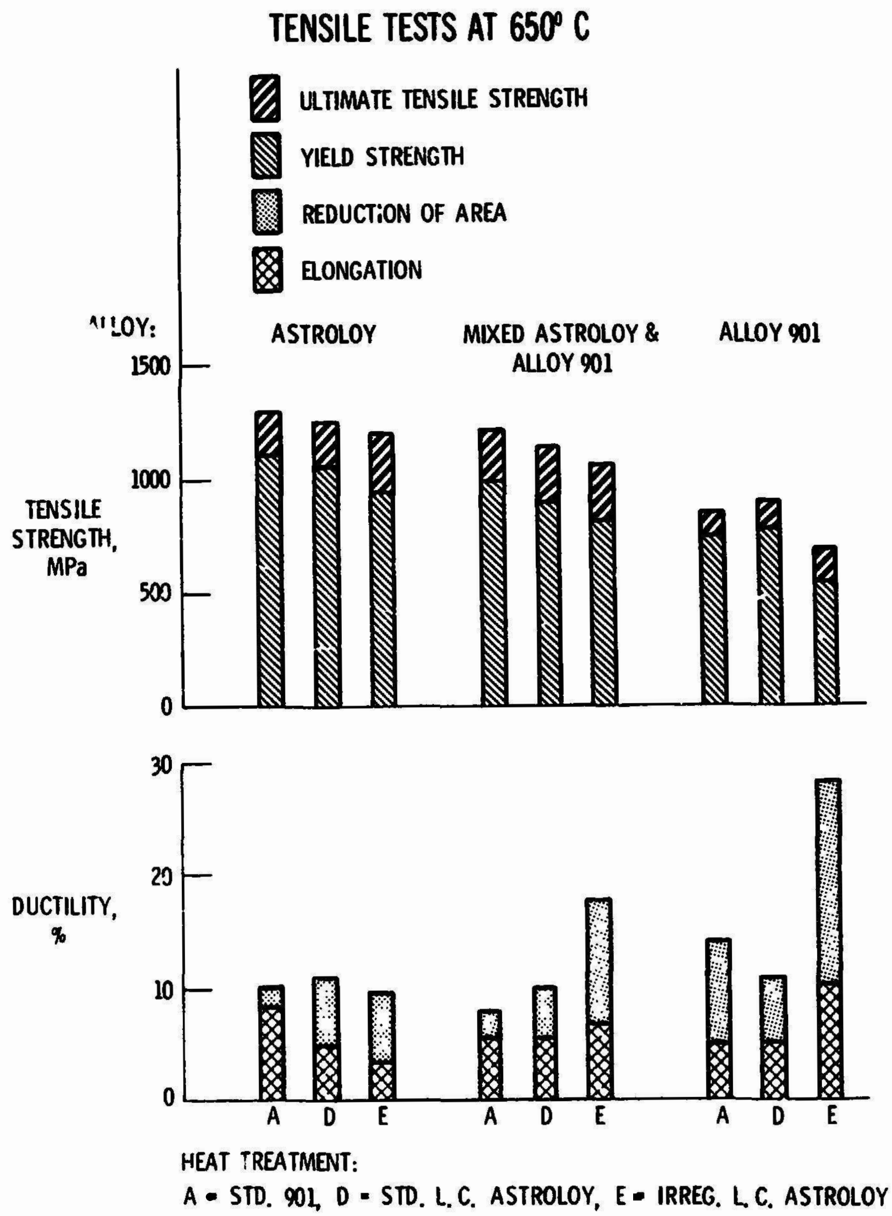
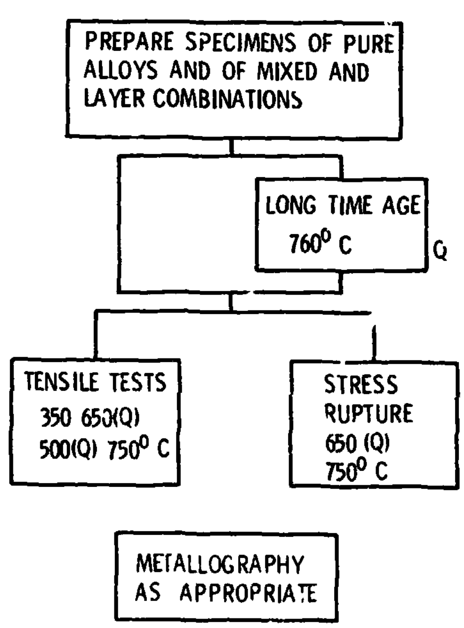


CONCLUSIONS

- MICROSTRUCTURAL EVIDENCE SUGGESTS THAT HOT ISOSTATIC PRESSING OF PREALLOYED POWDERS DEVELOPS A STRONG BOND AT INTERFACES BETWEEN THE HIGH IRON CONTENT SUPERALLOY 901 AND SUPERALLOY RENE 95 OR LOW CARBON ASTROLOY
- AT 650^{0} C THE TENSILE PROPERTIES OF MIXED POWDERS OF RENE 95 - ALLOY 901 AND OF LOW CARBON ASTROLOY - ALLOY 901 WERE INTERMEDIATE TO THE CONSTITUENT BASE ALLOYS
- BASED ON LIMITED DATA, THE JOINING BY A POWDER-METALLURGY/HIP TECHNIQUE OF SOME IRON-BASE AND NICKEL-BASE SUPERALLOYS APPEARS VIABLE

FUTURE WORK: DUAL ALLOY INTERFACE STABILITY

DETAILED PROPERTY DETERMINATIONS



N83 11298 D16

REDUCTION OF CHROMIUM IN Ni-BASE SUPERALLOYS THROUGH ELEMENT SUBSTITUTION AND RAPID SOLIDIFICATION PROCESSING

H. D. Fraser and B. C. Mudal
University of Illinois
Urbana-Champaign, Illinois

A study of the possible reduction in the use of Cr in Ni-base superalloys by the combined approaches of both elemental substitution and rapid solidification processing is proposed. The elements Si, Zr, Y and Hf have been chosen as potential partial substitutes for Cr in Waspaloy and IN 713LC since their separate addition to other alloys has previously been shown to result in enhanced oxidation resistance. The program will consist of three thrusts. First, the roles of Cr and these replacement elements in determining the microstructure and properties will be evaluated. Second, the elements Si, Zr, Y and Hf will be used as partial replacements for Cr in the base superalloys and these resultant alloys will be processed using rapid solidification techniques. Finally, the mechanical properties and oxidation resistance of the processed materials will be evaluated. In each section, emphasis will be placed on characterizing microstructure using state-of-the-art techniques (e.g. analytical transmission electron microscopy), and determining the mechanism by which these structures are produced.

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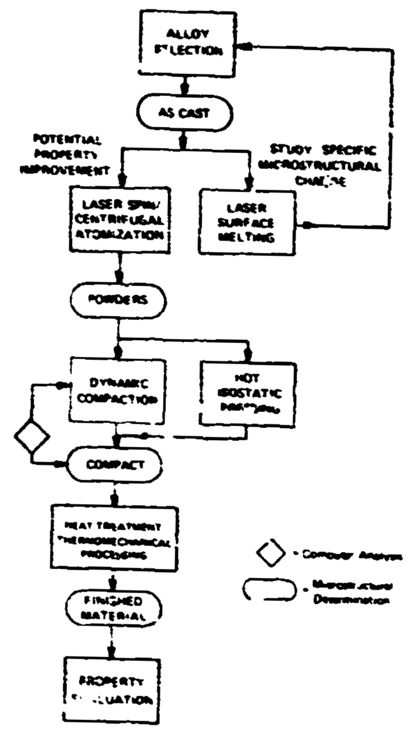
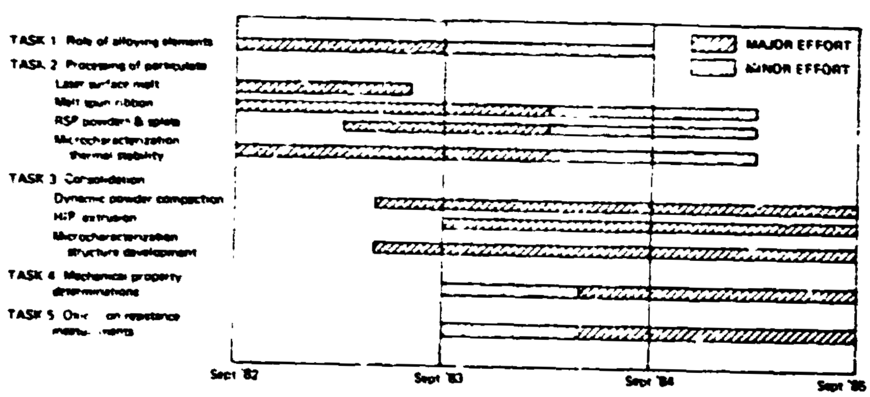


UNIVERSITY OF ILLINOIS



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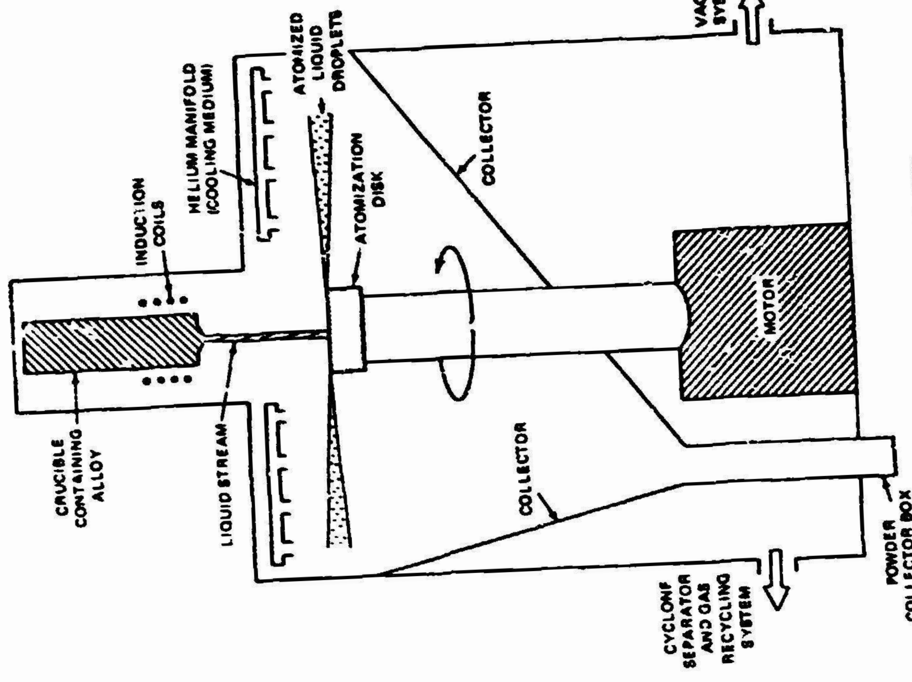
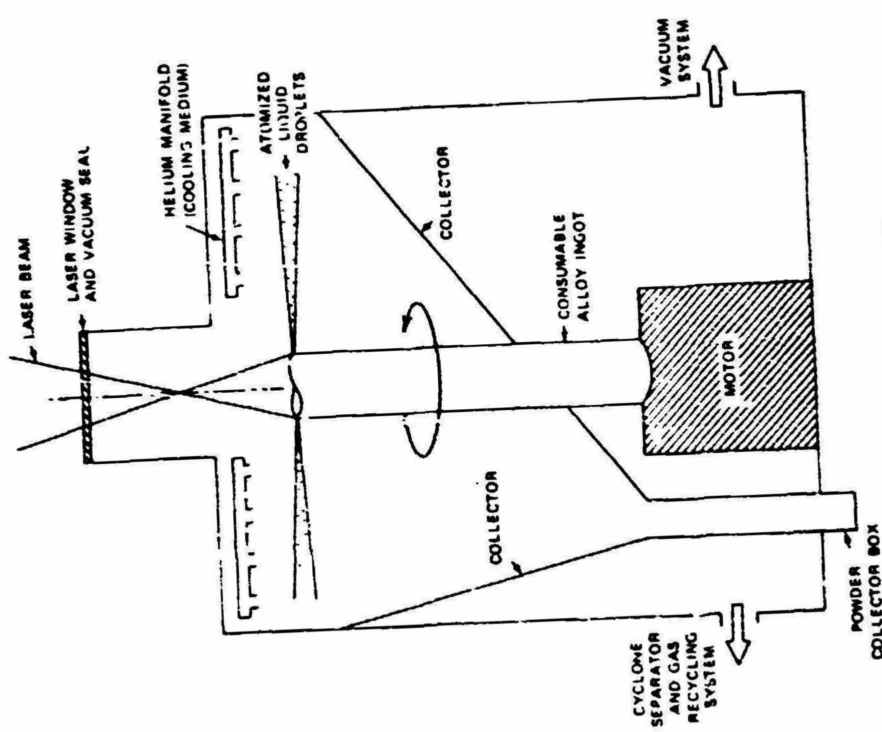
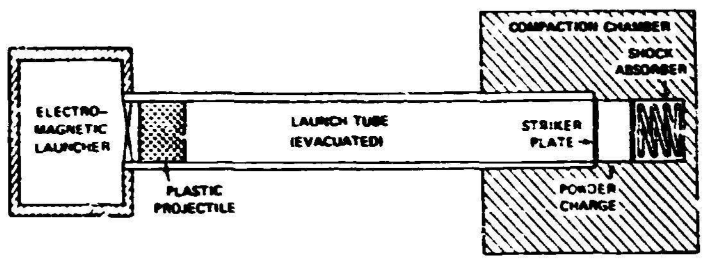


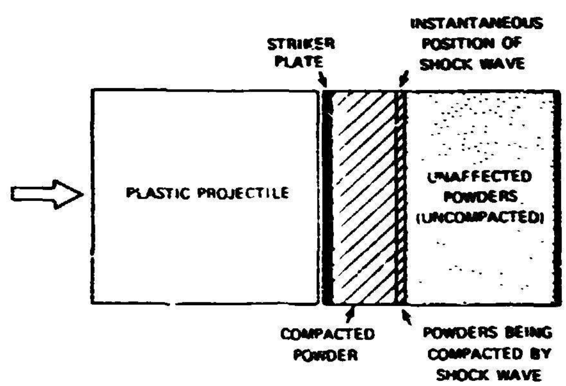
CENTRIFUGAL ATOMIZATION DEVICE



LASER SPIN ATOMIZATION DEVICE



DYNAMIC POWDER COMPACTION DEVICE



"DYNAMIC" NATURE OF PROCESS

N83 11299 D/7

EFFECTS ON STRESS RUPTURE LIFE AND TENSILE STRENGTH OF TIN ADDITIONS TO INCONEL 718

Robert L. Dreshfield
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

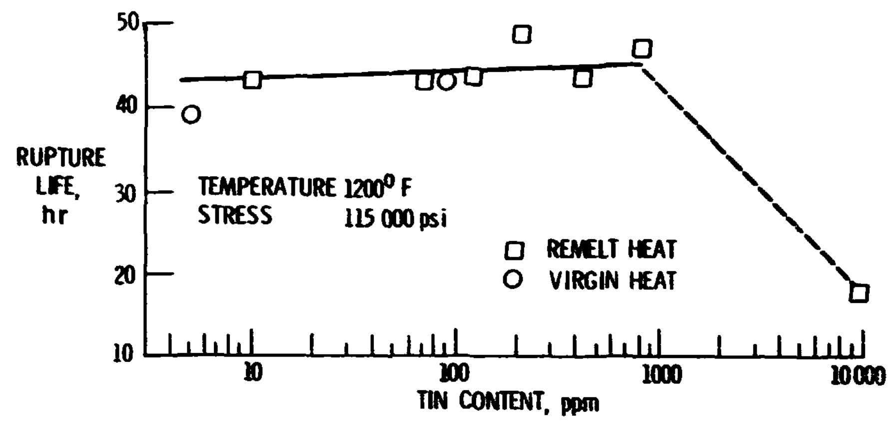
and

Walter Johnson
Special Metals Corporation
New Hartford, New York

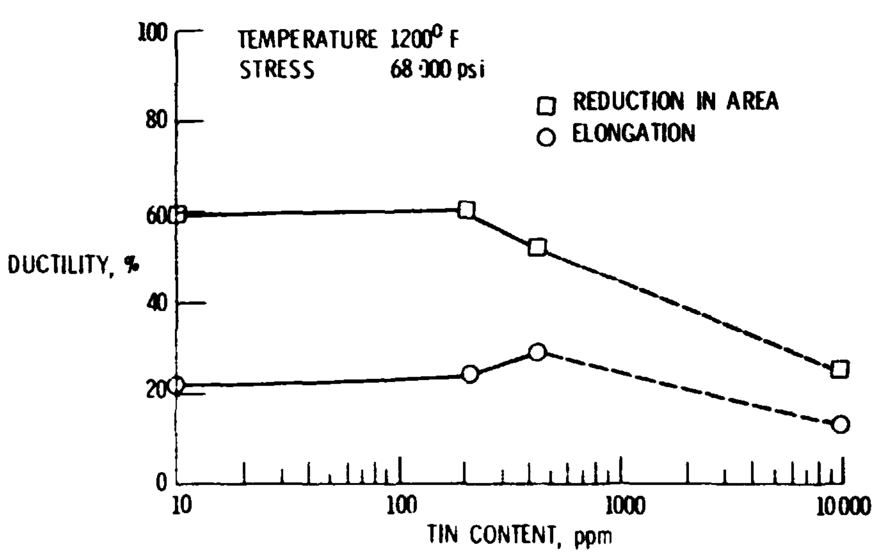
Columbium is ranked among the metals considered to be strategic metals for the United States aerospace industry. Because Inconel 718 represents a major use of columbium and a large potential source of columbium for aerospace alloys could be that of columbium derived from tin slags, this investigation was initiated to determine the effects of tin additions to Inconel 718 at levels which might be typical of or exceed those anticipated if tin slag derived columbium were used as a melting stock.

For this study, tin was added to 15 pound Inconel 718 heats at levels varying from none added to approximately 10 000 ppm (1 wt.%). Limited 1200° F stress rupture testing was performed at stresses from 68 000 to 115 000 psi and a few tensile tests were performed at room temperature, 800° and 1200° F. Additions of tin in excess of 800 ppm were shown to be detrimental to ductility and stress rupture life. The results of the investigation suggest that a more thorough study of the effects of tin on the mechanical properties of Inconel 718 is warranted to establish acceptable tin levels.

EFFECT OF TIN CONTENT ON RUPTURE LIFE OF INCONEL 718



EFFECT OF TIN ON STRESS RUPTURE DUCTILITY OF INCONEL 718



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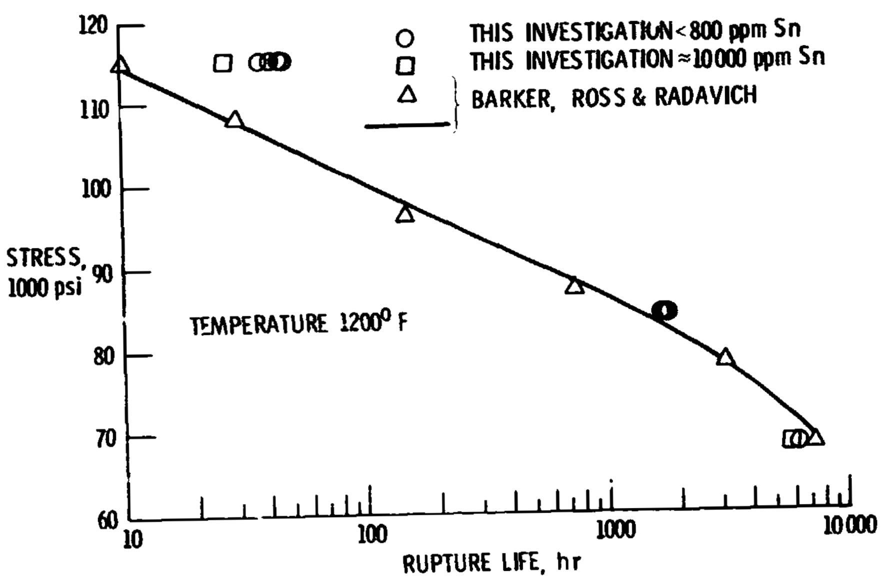
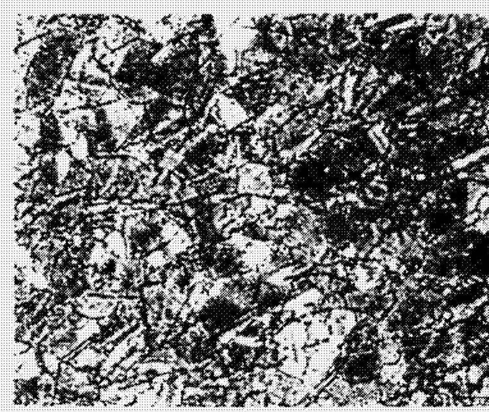
STRESS RUPTURE LIFE OF INCONEL 718

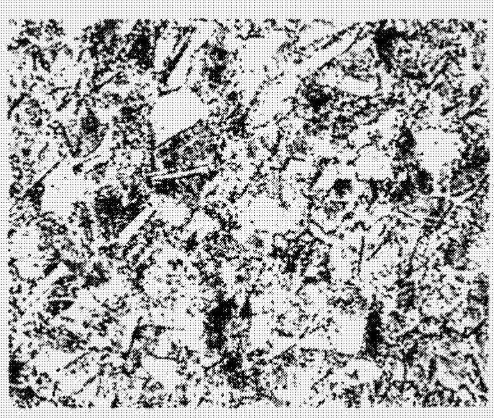


INCONEL 718 AFTER STRESS RUPTURE TEST AT 1200° F

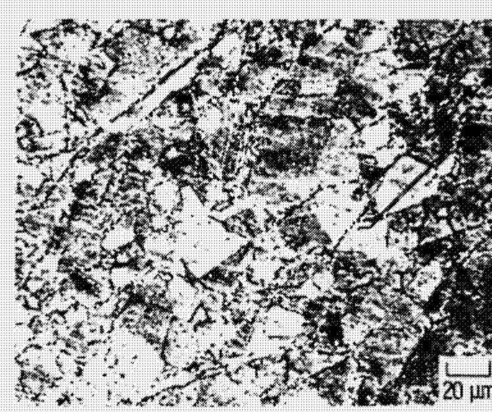
10 ppm TIN
5752 hr



210 ppm TIN
6087 hr



420 ppm TIN
6189 hr



9900 ppm TIN
5839 hr

ETCH: 35 ml ETHANAL, 65 ml HYDROCHLORIC ACID
7 DROPS HYDROGEN PEROXIDE

CS-82-2168

INCONEL 718 WITH 9900 ppm TIN

800° F TENSILE TEST

CS-82-2169

OMIT

COMPILATION AND CRITICAL EVALUATION OF NICKEL BINARY PHASE DIAGRAMS

Philip Nash
Illinois Institute of Technology
Department of Metallurgical and Materials Engineering
Chicago, Illinois 60616

The American Society for Metals and the National Bureau of Standards have initiated an alloy phase diagram data program of which one of the principal associates is the NASA - Lewis Research Center. This program aims to compile and critically evaluate all of the published data on binary phase diagrams and this work will be of great value to the COSAM program.

The essential elements of the method and techniques used to compile and critically evaluate the Nickel binary phase diagrams will be described. The published literature on each system is compiled from a computer search of the Metadex and Chemical abstracts files supplied by ASM and hardcopies of pertinent references obtained. All the data pertaining to the graphical representation of the diagram are input to a computer via a graphics tablet so that the data may be stored and manipulated for ease of comparison between different investigations. The bibliography, text and other data are also stored on disc for ease of access and manipulation. A least squares optimization program is used in conjunction with available thermodynamic and phase equilibrium data to produce a consistent phase diagram. The evaluation of each system will include metastable phase equilibria in addition to stable phase equilibria.

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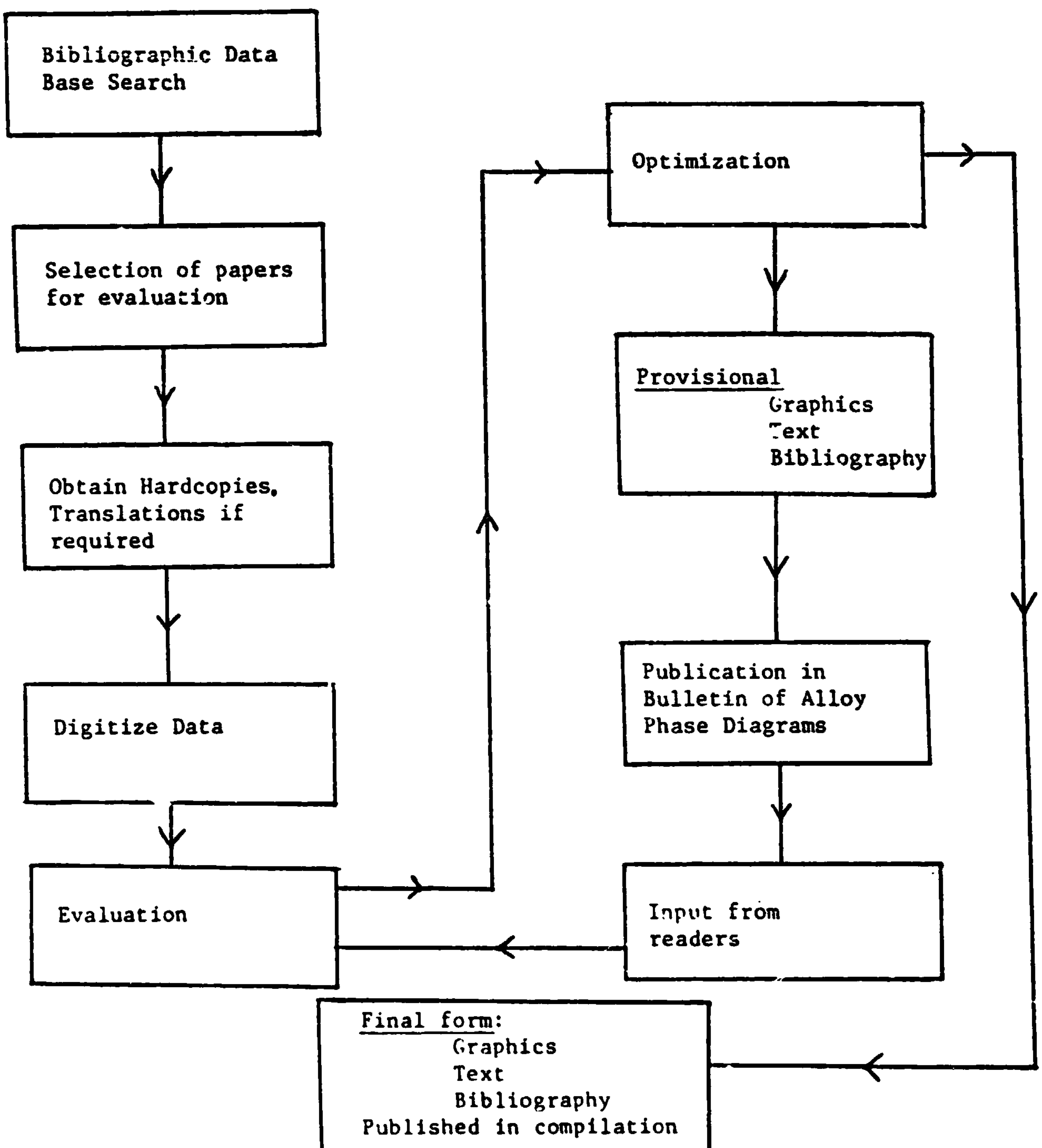
Phase Diagram Evaluation Procedure



N83 11300 D/8

INTERMETALLICS AS ALTERNATIVE MATERIALS

J. Daniel Whittenberger
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

Intermetallics represent a large class of materials whose potential for use in high temperature aggressive environments has basically been ignored in favor of ceramics. While many candidate intermetallics are brittle, directionally bonded compounds; a few, such as the equiatomic aluminides of iron, nickel, and cobalt, do possess metallic-like behavior. These aluminum containing intermetallics have B2 cubic crystal structures, exist over a wide range of composition, have large solubilities for third element alloying additions, are capable of both low and high temperature plastic flow, and have very high melting temperatures (~1900 K) except for FeAl.

A program has been initiated at the Lewis Research Center to investigate the slow strain rate elevated temperature properties of Fe, Ni, and Co aluminides. Because of the reported difficulties with traditional melting/casting methods, sound polycrystalling materials are currently being fabricated by hot extrusion of steel canned blended prealloyed powders. These binary aluminides are being used in both in house studies and grant programs to develop base line elevated temperature mechanical properties as well as an understanding of the factors which affect/control the strength and ductility.

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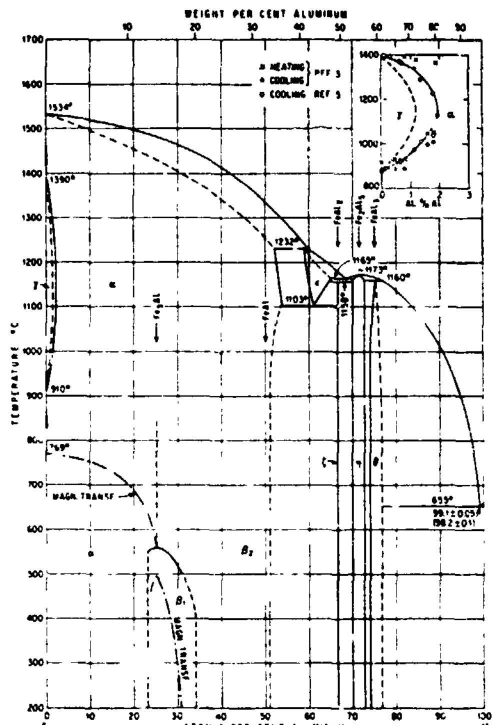
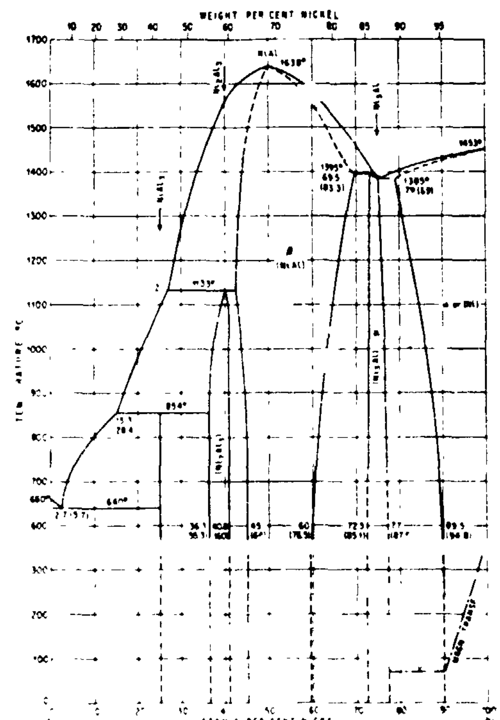
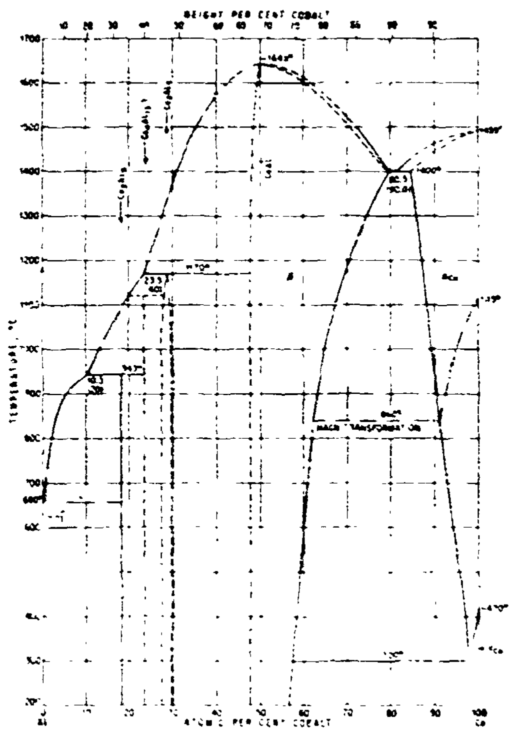
IRON-ALUMINUM PHASE DIAGRAM



ALUMINUM-NICKEL PHASE DIAGRAM



ALUMINUM-COBALT PHASE DIAGRAM



WHY THESE ALUMINIDES?

1. CUBIC (B2) CRYSTAL STRUCTURES

2. BINARY ALUMINIDES EXIST OVER A WIDE RANGE IN COMPOSITION AND HAVE A LARGE SOLUBILITY FOR SUBSTITUTIONAL 3rd ELEMENT ADDITIONS

3. CoAl AND NiAl HAVE VERY HIGH MELTING POINTS (→ 1900 K) FeAl HAS A LOWER MELTING POINT (> 1500 K) BUT CONTAINS INEXPENSIVE, READILY AVAILABLE ELEMENTS

4. POSSESS POTENTIAL FOR SELF PROTECTION IN OXIDIZING ATMOSPHERE

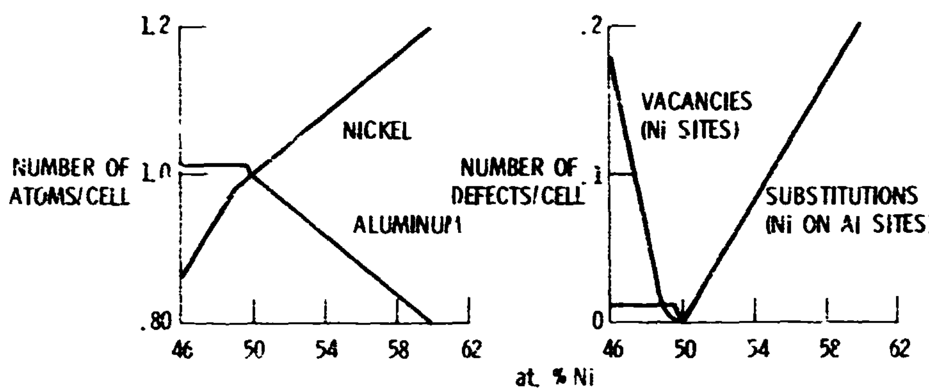
DIFFERENCES WITH RESPECT TO ALLOYS

- ORDERED CRYSTAL STRUCTURE

- HIGH POINT DEFECT CONCENTRATIONS

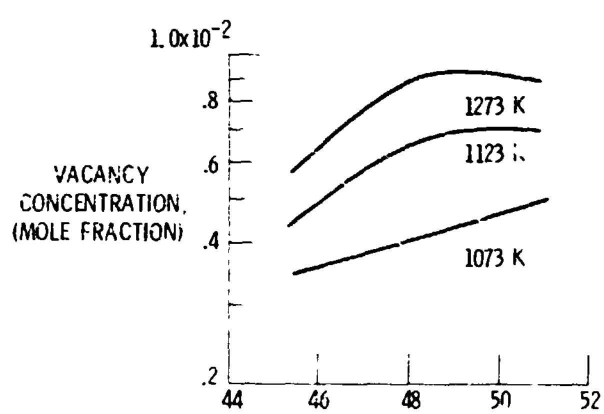
DEFECT STRUCTURE IN NiAl

(A. J. BRADLEY AND A. TAYLOR: PROC ROY SOC., A 159 (1937), 56)

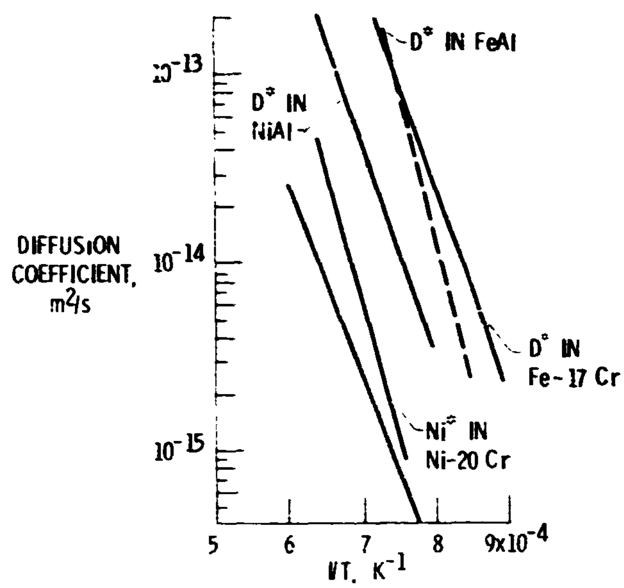


THERMAL VACANCY CONCENTRATION IN FeAl

(K. HO AND R.A. DODD: SCRIPTA MET., 12 (1978), 1055)

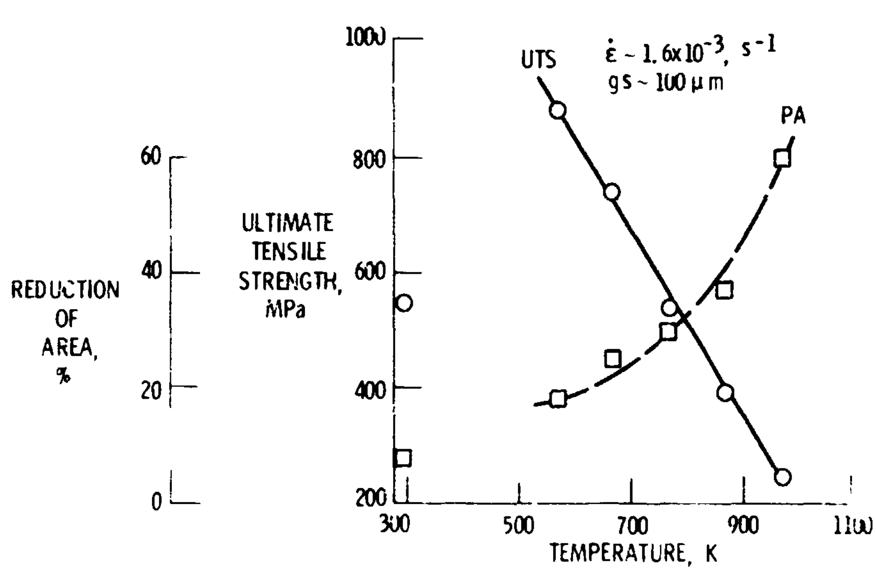


CATION TRACER DIFFUSION COEFFICIENTS IN SEVERAL ALUMINIDES AND ALLOYS



TENSILE PROPERTIES OF Fe-40Al

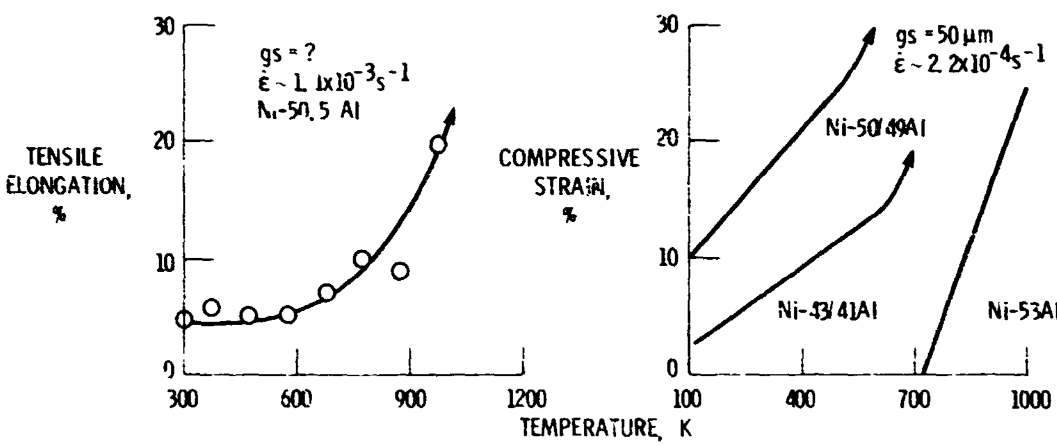
(G. SAINFORT, et al: "FRAGILITE et EFFECTS de L'IRRADITION," PRESSES UNIVERSITAIRES de FRANCE, PARIS, FRANCE, 1967 pp 187-98)



DUCTILITY OF POLYCRYSTALLINE NiAl

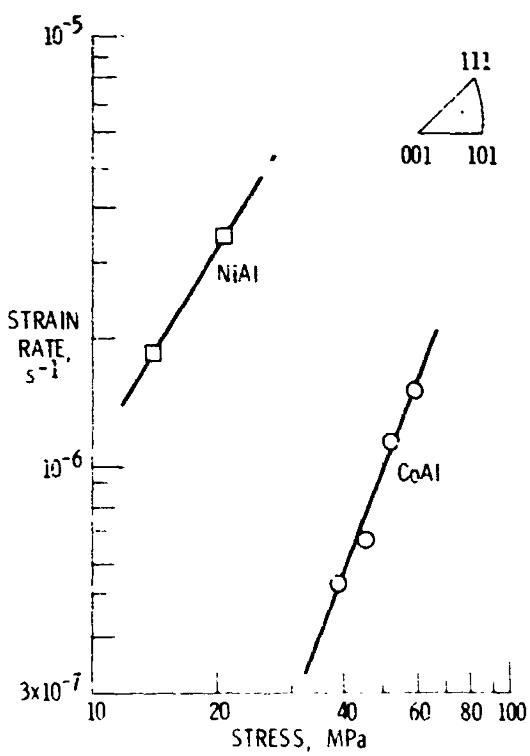
(A.G. ROZNER AND R.J. WASILEWSKI: J.I.M., 94 (1966), 169)

(R.T. PASCOE AND C.W.A. NEWEY: MET SCI. J., 2(1968) 138)



COMPRESSIVE CREEP STRENGTH OF Co-50Al AND Ni-50Al SINGLE CRYSTALS AT 1323 K

(L.A. HOCKING, P.R. STRUTT AND R.A. DODD: J.I.M. 99 (1971), 98-101)



DISLOCATION STRUCTURE

NiAl - BETWEEN 1964 AND 1974 CONSIDERABLE EFFORT ON LARGE GRAIN SIZE POLYCRSTALLINE MATERIALS AND SINGLE CRYSTALS AFTER RELATIVELY FAST STRAIN RATE ($\sim 10^{-6}$ sec^{-1}) TESTING

- BURGER'S VECTORS <001>
  <011>
  <111>
- SUBGRAINS FORMED DURING DEFORMATION EXCEPT FOR Ni-45Al ALLOY

FeAl - SOME EFFORT ON SINGLE CRYSTALS

- BURGER'S VECTORS <111>
  <100>

CoAl - ONLY ONE EXPERIMENT

- BURGER'S VECTOR <100> PROBABLE
- NO DISLOCATION SUBSTRUCTURE FOUND

OBJECTIVE

MEASURE THE SLOW PLASTIC PROPERTIES OF CoAl, FeAl, AND NiAl IN AIR AND DETERMINE THE MECHANISM(S) WHICH AFFECT THESE PROPERTIES

APPROACH

UNDERSTAND SLOW PLASTIC BEHAVIOR IN TERMS OF EXISTING DEFORMATION MODELS AND STRUCTURAL PARAMETERS

MODEL

$$\dot{\varepsilon} \propto \frac{1}{b} \left(\frac{E}{T}\right) D_{eff} \left(\frac{\sigma}{E}\right)^{n}$$

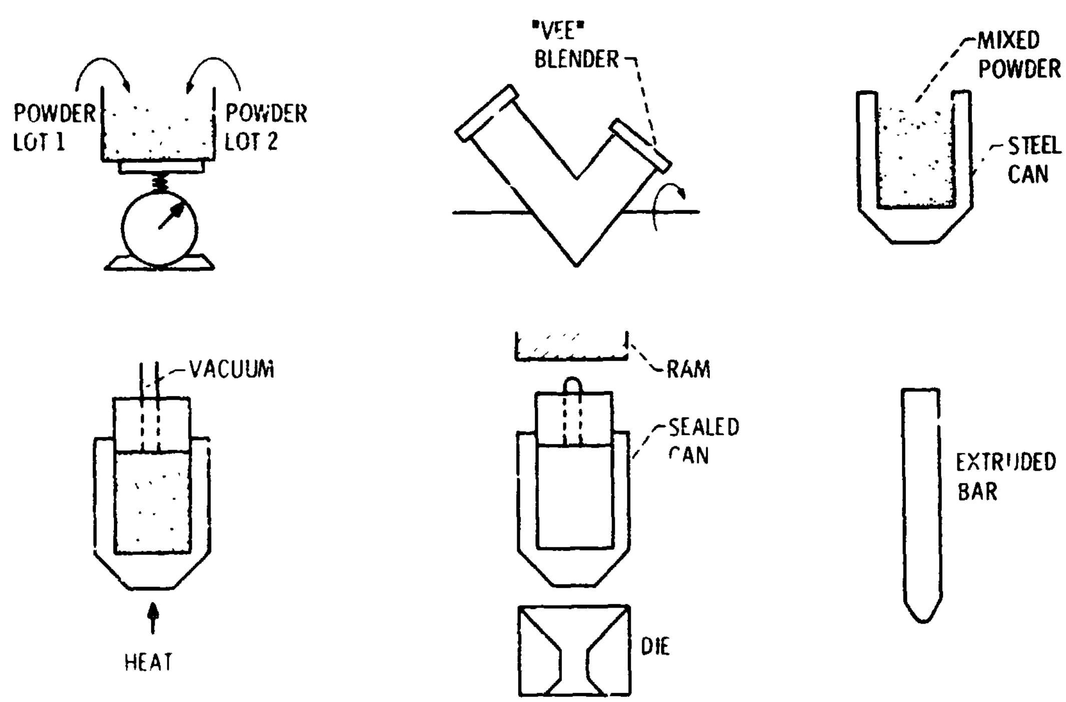
STRUCTURAL PARAMETERS

1. TYPES OF DISLOCATIONS, FAULTING, AND SUBSTRUCTURE
2. CRYSTAL ORIENTATION
3. GRAIN SIZE
4. COMPOSITION (CONCENTRATION AND TYPE(S) OF POINT DEFECTS)
5. GRAIN BOUNDARY BEHAVIOR

STATUS

1. SLOW PLASTIC STRAIN RATE TESTING	• COMPRESSIVE FLOW STRESS AS FUNCTIONS OF COMPOSITION, TEMPERATURE, STRAIN RATE, GRAIN SIZE
2. DISLOCATION STRUCTURE	• FeAl - R. V. KRISHNAN (NSF FELLOW) NiAl AND CoAl - W. D. NIX AND R. SINCLAIR (STANFORD UNIVERSITY)
3. DYNAMIC ELASTIC MODULUS	• A. WOLFENDEN (TEXAS A AND M UNIVERSITY)
4. VACANCY CONCENTRATION	• THERMAL EXPANSION LATTICE PARAMETER MEASUREMENTS
5. LOW TEMPERATURE DUCTILITY	• E. SCHULSON (DARTMOUTH COLLEGE)

SCHEMATIC OUTLINE OF POWDER METALLURGICAL TECHNIQUES UTILIZED TO PRODUCE INTERMETALLIC ALLOYS



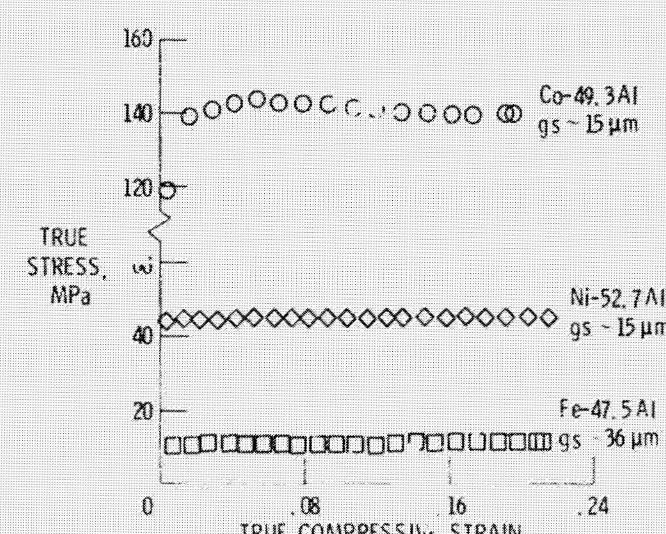
TYPICAL PHOTOMICROGRAPHS OF INTERMETALLIC ALLOYS

THE EXTRUSION AXIS IS VERTICAL

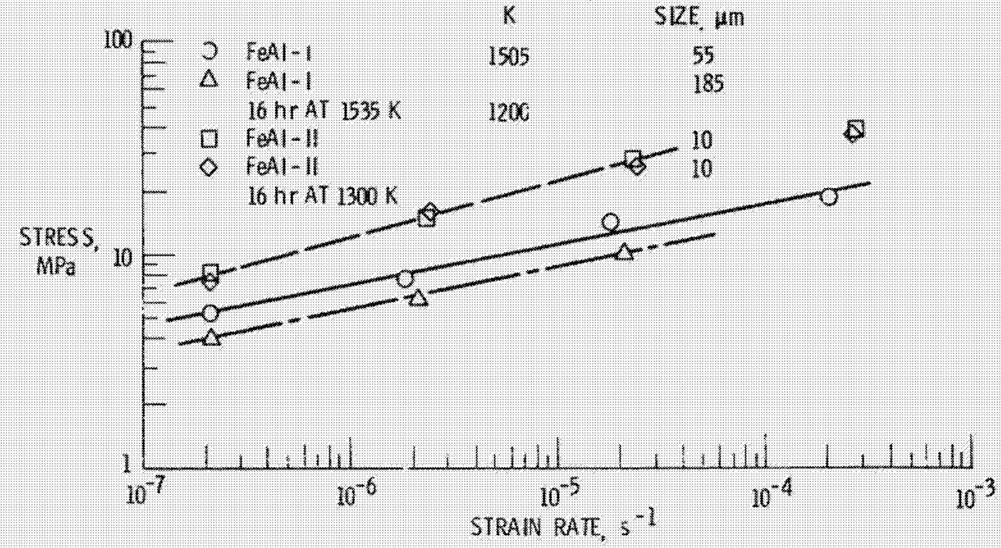
Co - 51.3 Al | Fe - 41.7 Al | Ni - 48.3 Al

CS-82-2252

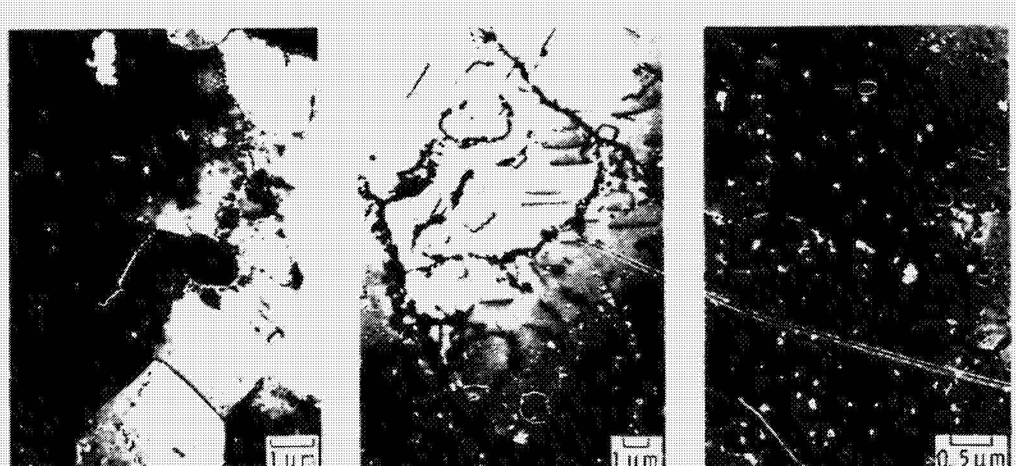
FLOW BEHAVIOR OF SEVERAL B2 ALUMINIDES AT 1300 K AND $\dot{\varepsilon} \approx 1.8 \times 10^{-4}$ -1



FLOW STRESS - STRAIN RATE BEHAVIOR FOR SEVERAL Fe-40Al MATERIALS AT 1200 K



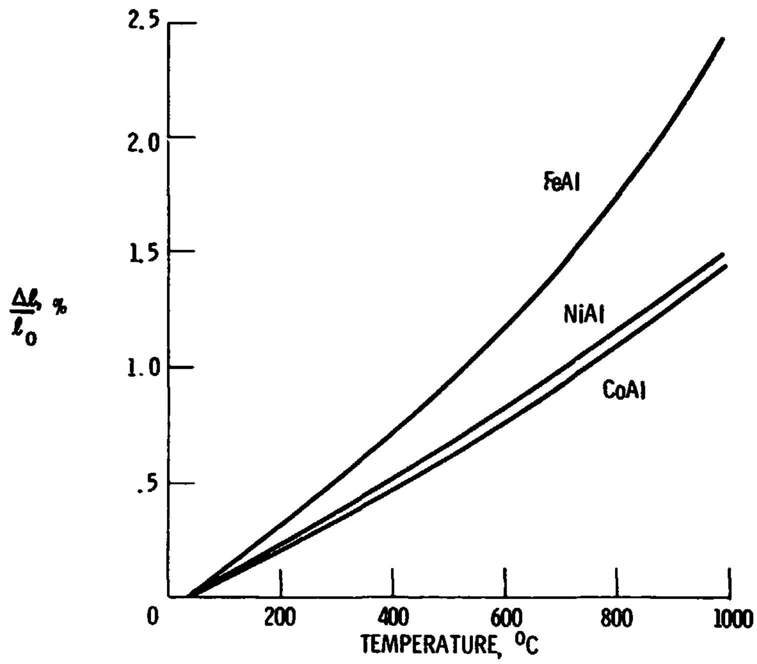
TYPICAL TEM PHOTOMICROGRAPHS OF AS EXTRUDED AND TESTED Fe-39.8 Al



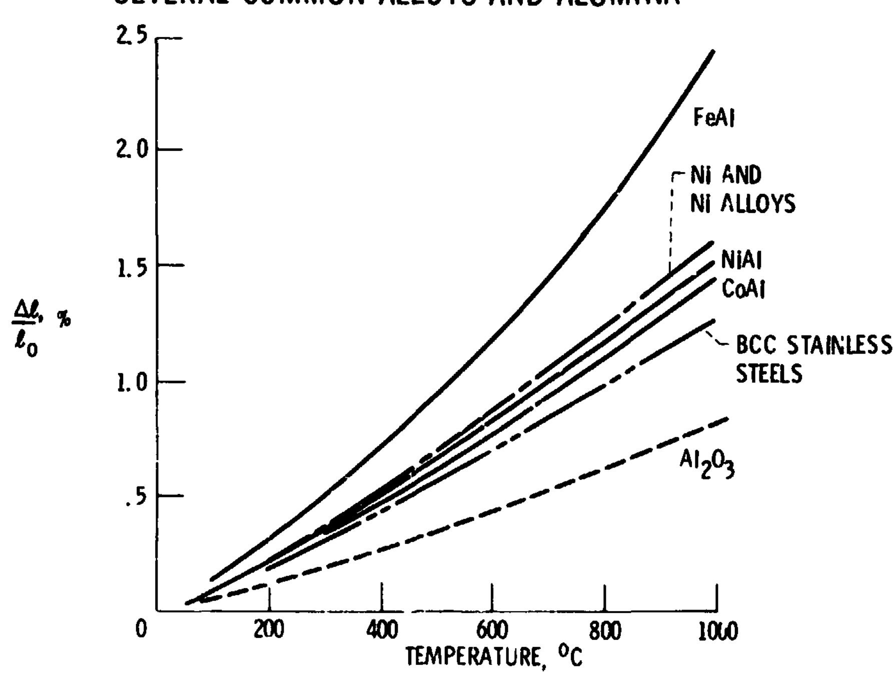
AS EXTRUDED 1200 K 16,1	TESTED 1200 K - $\dot{\varepsilon}$ 2x10^{-4} s^{-1} TO 28.5 % STRAIN	TESTED 1200 K - $\dot{\varepsilon}$ 2x10^{-7} s^{-1} TO 4.7 % STRAIN

CS-82-2251

THERMAL EXPANSION OF SEVERAL ALUMINIDES



COMPARISON OF THE THERMAL EXPANSION OF ALUMINIDES
SEVERAL COMMON ALLOYS AND ALUMINA



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FUTURE WORK

- PLASTIC FLOW BEHAVIOR
- TEM
- MODULUS
- INITIATE EFFORTS ON DIFFUSION AND POSSIBLY GRAIN BOUNDARIES
- THIRD ELEMENT ALLOYING

RESULTS TO DATE

- POLYCRYSTALLINE ALUMINIDES CAN BE FABRICATED VIA POWDER METALLURGY TECHNIQUES
- POLYCRYSTALLINE CoAl IS STRONG AND DUCTILE
- GRAIN SIZE STRENGTHENING FOR $T/T_M \leq 0.75$
- POSSIBLE PROBLEM WITH OXIDATION RESISTANCE OF FeAl

N83 11301 D/9

THE STRENGTH AND DUCTILITY OF POLYCRYSTALLINE NiAl IN TENSION

E. M. Schulson
Thayer School of Engineering
Dartmouth College
Hanover, New Hampshire 03755

The purpose of this paper is to review the results of an experimental study under way at Dartmouth on the tensile strength and ductility of the B2 aluminide, NiAl. Specifically, ductility at low temperatures is being sought through two routes, grain refinement and microalloying. Experiments at temperatures from 20°C to 400°C at two strain rates (1×10^{-4} S^{-1} and 5×10^{-6} S^{-1}) have established that.

i) at room temperature, binary and microalloyed (< 1000 ppm La, Y, Mo, Ti) NiAl shows negligible ductility, independent of grain size over the range 5 to 140 µm;

ii) at 295°C the tensile elongation of binary 51 Ni/49 Al increases from <1% to about 5% upon decreasing the grain size to below ≈ 10 µm;

iii) similarly, at 400°C the ductility increases from about 2% to > 15% upon decreasing the grain size to below 15 µm;

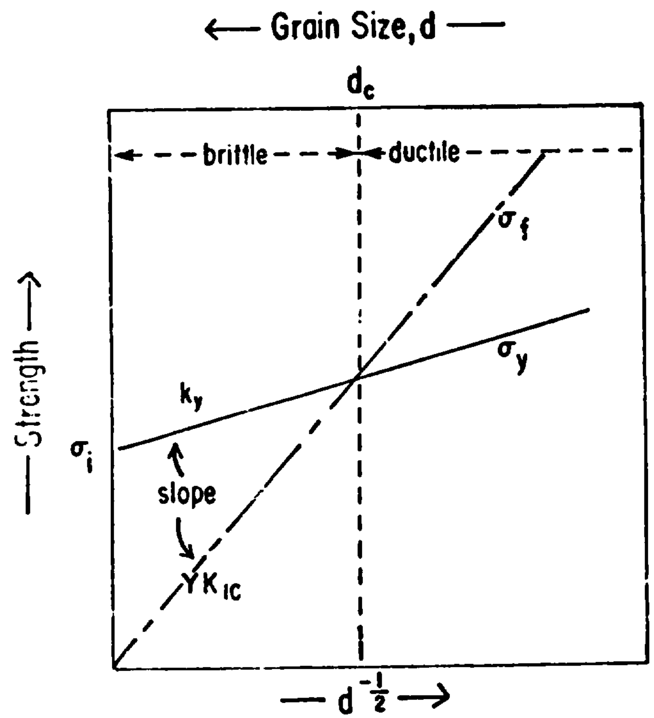
iv) the ductility of fine-grained (7 µm) binary aggregates deformed at 295°C increases from ≈ 5% to 12% upon decreasing the strain rate from 10^{-4} S^{-1} to 5×10^{-6} S^{-1};

v) partial recrystallization (10 to 20%) of warm-extruded binary and microalloyed material imparts 1 to 2% ductility at room temperature where fully recrystallized material is brittle (point (i));

vi) the yield strength obeys a Hall-Petch relationship; and

vii) when ductility is not observed, fracture coincides with yielding.

The mechanisms underlying the flow and fracture of NiAl are discussed in terms of the nucleation and growth of microcracks. The concept of a critical grain size, presented elsewhere (E.M. Schulson, Res. Mech. Lett. 1 (1981) 111), is considered in the light of the above results.



$$d_c = \left(\frac{YK_{IC} - k_y}{\sigma_i} \right)^2$$

TEST MATERIALS

Alloy	Composition (wt.%)						
	Ni	Al	Mo	Ti	Y	La	B
1*	69.3	30.7					
2	69.2	30.7	0.075				
3	69.3	30.7				< 0.33	
4	69.3	30.7			0.02		
5	69.2	30.7	0.058	0.056			
6	69.2	30.7					0.10

*The base, Alloy-1 is 51 at.% Ni/ 49 at.% Al

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TRACE ELEMENTS (ppm) IN ALLOY-1

Element	Concentration (ppm) in Alloy-1
C	20
O	< 70
N	3
S	< 10
Zr	125
Fe	200
Si	100
Mg	10

MATERIALS PROCESSING

1) Hot-extrude* ingots to 19 mm rod at ≈ 1000°C through area reduction ratio of 7:1.

2) Re-extrude* to 6 mm rod at ≈ 550°C through area reduction ratio of 7:1.

3) Recrystallize+ at temperatures between 700°C and 800°C to produce either partially recrystallized or fully recrystallized material of grain size from 5 µm to 140 µm.

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*See attached table of extrusion constants for details.

+See attached paper on recrystallization and grain growth.

Extrusion Conditions and Extrusion Constants for Binary and for Alloyed NiAl

Extrusion No.	Ingot & Alloy	Ext. Temp. (°C)	Avg. Ext. Speed (in./min)	Ext. Ratio, R	% of NiAl in Billet	F_B [(a)] (tons)	K_B [(b)] (ksi)	F_F [(c)] or (tons)	F_{SS} [(d)] (tons)	K_F or K_{SS} (ksi)
	-- Hot Extrusions of Ingots to 0.75 in. rod --									
80-36	Binary	1090	-	8.0	100	211	58		172	47
81-25	Binary	997	16	7.9	82	201	55	261		72
81-32	Binary	1000	16	7.0	82	191	56	264		77
81-33	+B	992	stalled	7.0	82	300	88	-press stalled at 347 tons -		
81-34	+Mo	1008	26	6.9	82	218	64	234		69
81-40	+Mo+Ti	1000	26	6.9	82	195	57	218		65
81-46	+La	1005	23	6.9	82	191	56	234		59
81-47	+Y	1002	24	6.9	82	195	57	224		66
81-54	Binary	955	31	6.9	83	231	68		191	56
81-55	+B	1070	15	6.9	82	284	83	327		96
	-- Warm Re-extrusions --- 0.75 in. rod to 0.25 in. rod --									
80-54	Binary (80-36)	499	stalled	7.9	11	press stalled after breakthrough				
81-52	Binary (81-32)	546	28	7.1	13	330	96		300	87
81-68	Binary (81-32)	474	37	4.5	13	[illegible]	126		300	113
81-69	Binary (81-32)	565	37	7.1	13	327	95		281	82
81-70	Binary (81-32)	563	36	7.1	13	330	96		284	82

(a) F_B = breakthrough force

(b) K_B = extrusion constant at breakthrough

$= \frac{F_B/A}{\ln R}$ where A = cross-sectional area of billet

(c) F_F = final force

(c) F_{SS} = steady state force

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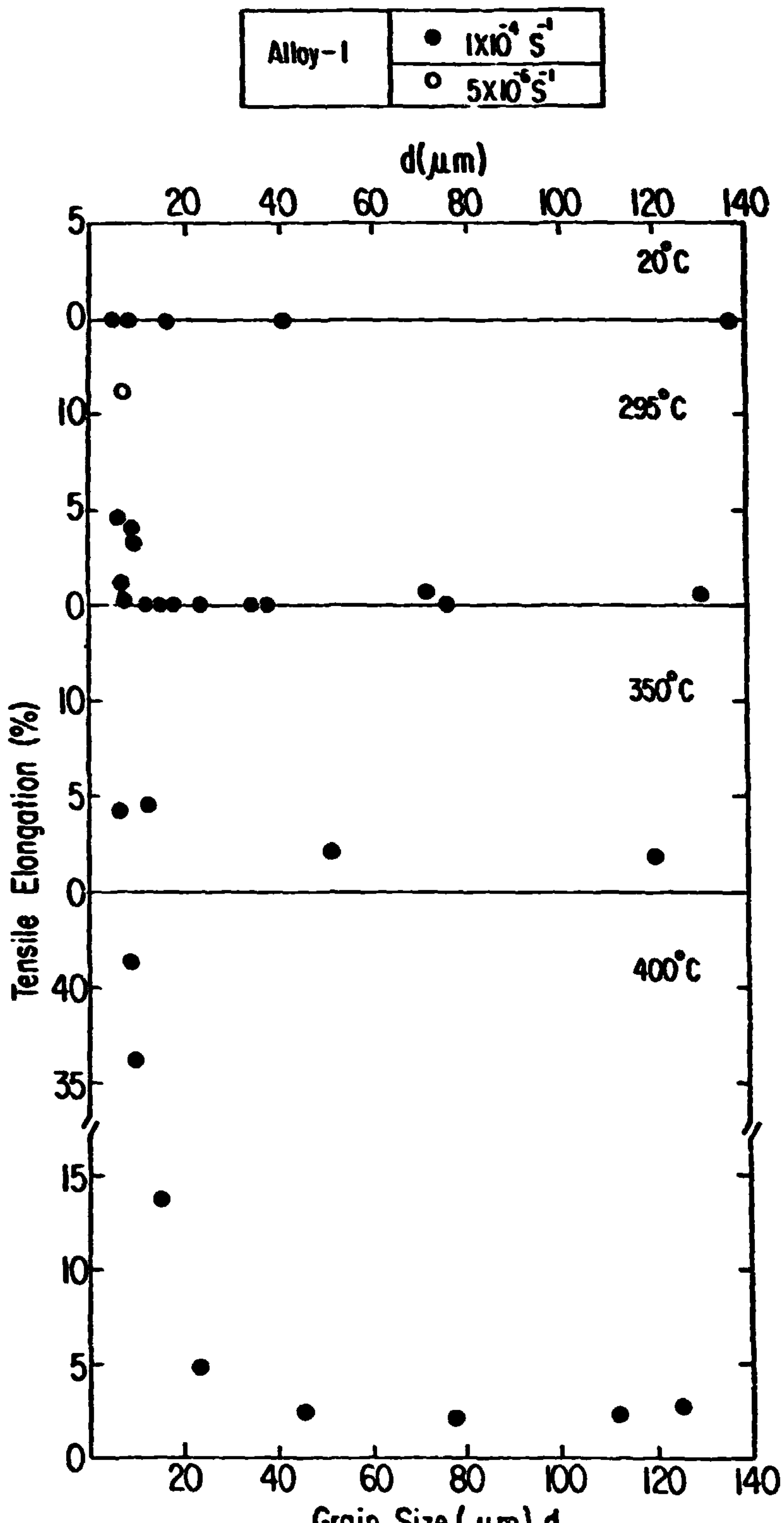

c-3

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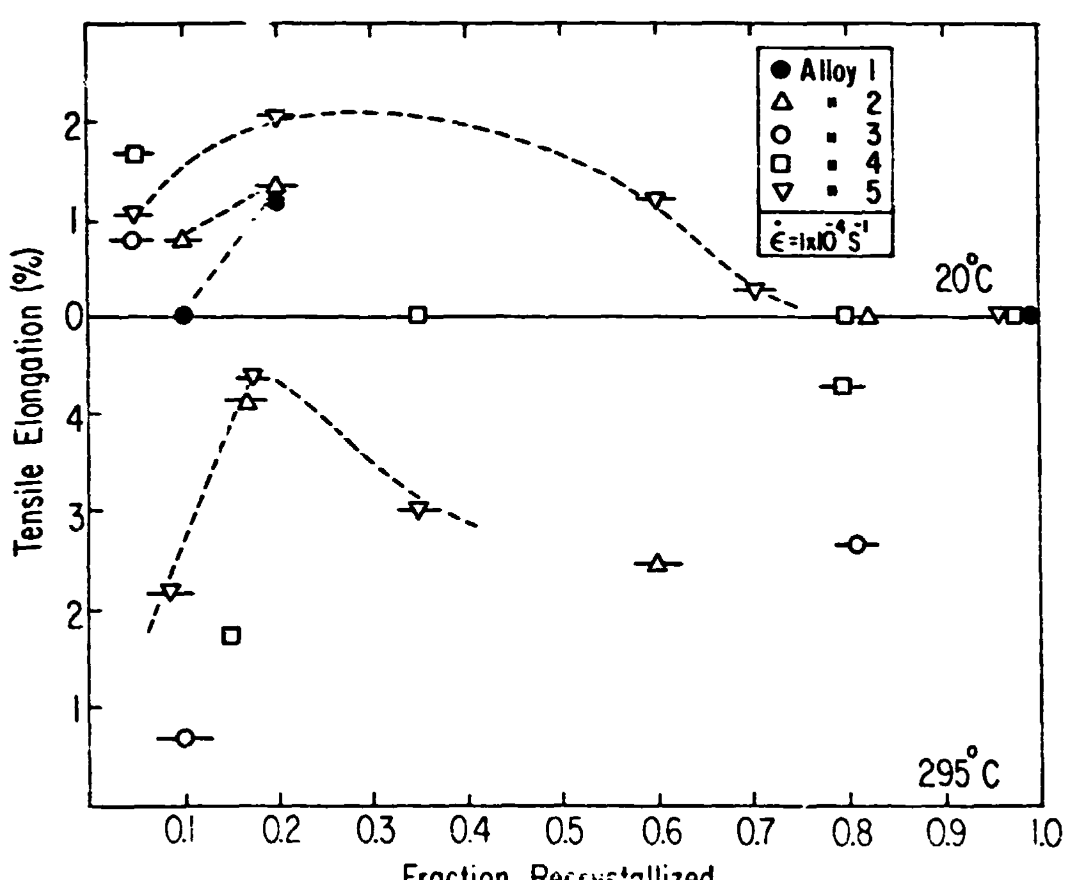
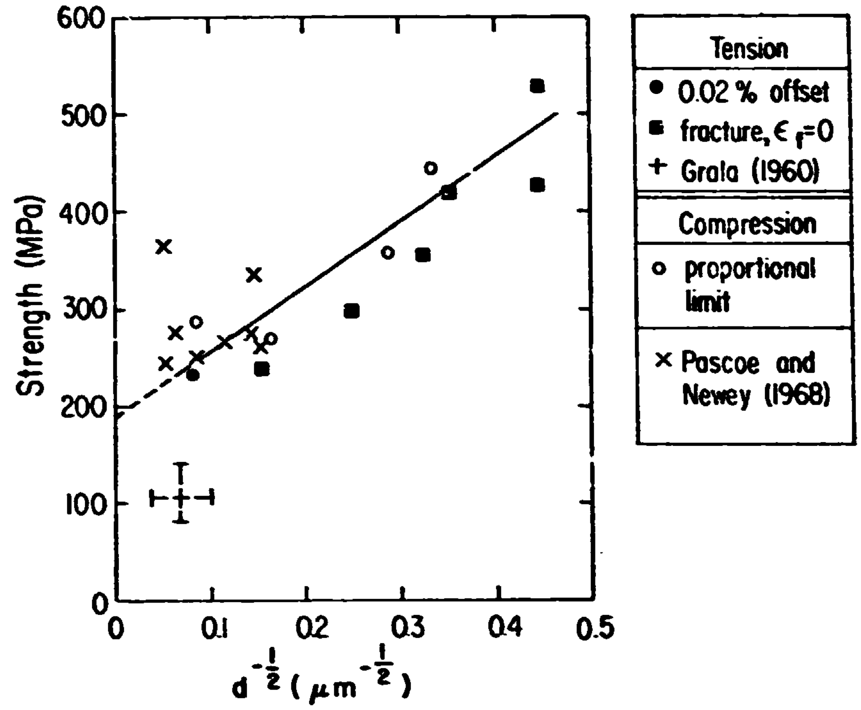
DUCTILITY OF PARTIALLY RECRYSTALLIZED NiAl

Material	Strain Rate (S^{-1})	Temp. (°C)	Fractional Recrystallization	Elongation (%)
Alloy-1	1×10^{-4}	20	0.2	1.2
		295	0.1	7.4
		400	0.1	64.1

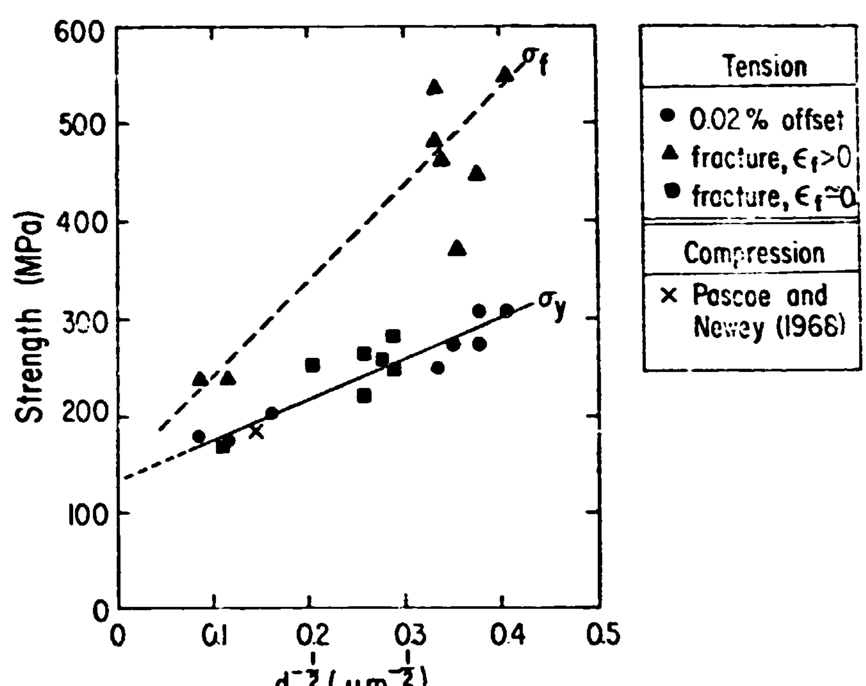


Alloy-I; 20° C



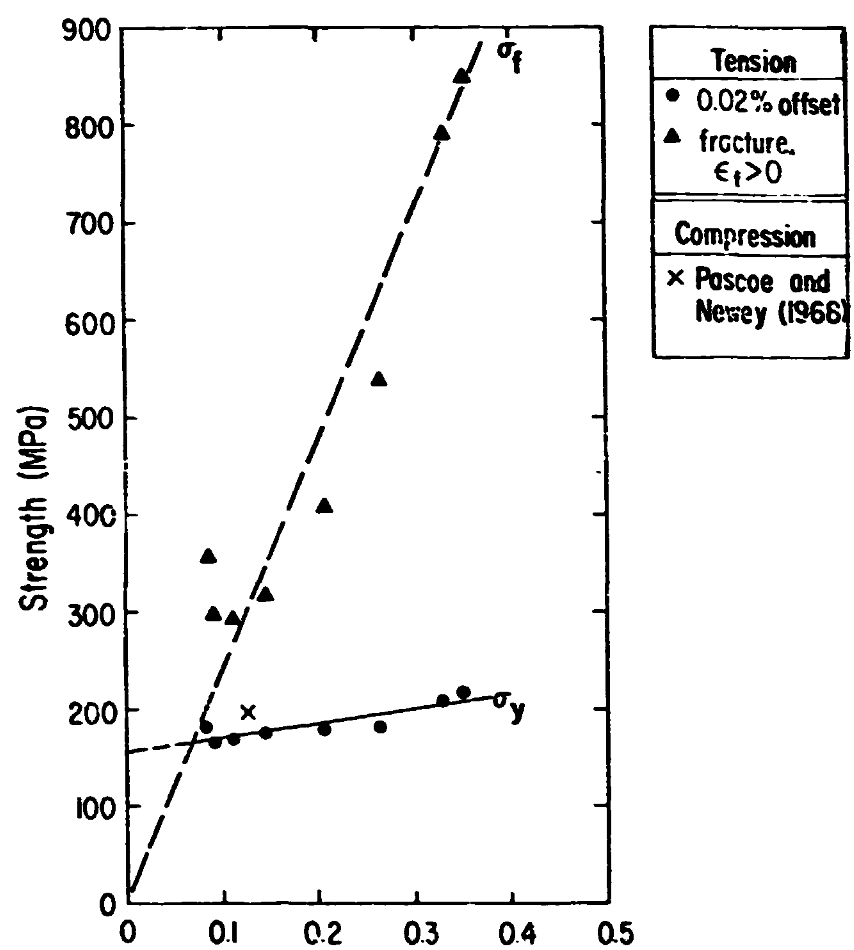
σy (MPa) = 189 + 0.67 d(m)$^{-1/2}$

Alloy-I; 295° C



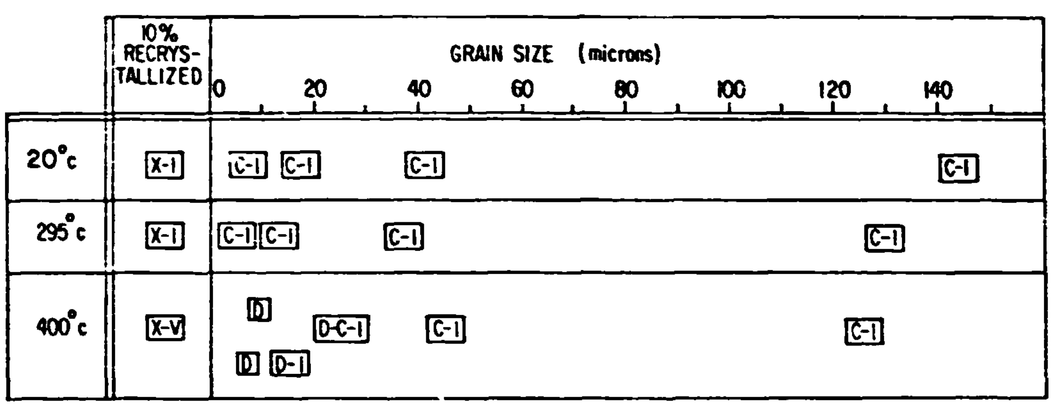
σy (MPa) = 133 + 0.41 d (m)$^{-\frac{1}{2}}$

Alloy -I; 400° C



σ_y (MPa) = 152 + 0.16 d(m)$^{-\frac{1}{2}}$

FRACTURE MODES: Alloy-I



i = Intergranular
C = Cleavage
D = Ductile, torn appearance
V = Microvoids
X = Unrecrystallized grains prevent characterization

N83 11302 D20

HIGH TEMPERATURE DEFORMATION OF NiAl AND CoAl

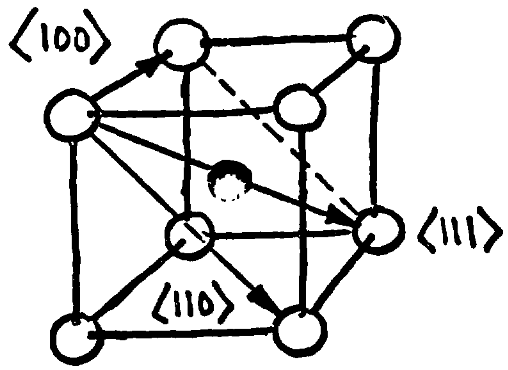
W. D. Nix
Stanford University
Department of Materials Science and Engineering
Stanford, California 94305

The high temperature mechanical properties of the aluminides are reviewed with respect to their potential as high temperature structural materials. It is shown that NiAl and CoAl are substantially stronger than the pure metals Ni and Co at high temperatures and approach the strength of some superalloys, particularly when those superalloys are tested in "weak" directions. The objective of the research in progress is to determine the factors that limit and control the high temperature strengths of NiAl and CoAl to provide a basis for the development of intermetallic alloys of this type.

The study of CoAl is motivated primarily by the observation that it is much stronger than NiAl, even though their structures and melting temperatures are the same. An understanding of this effect could lead to the replacement of Co with less strategic elements.

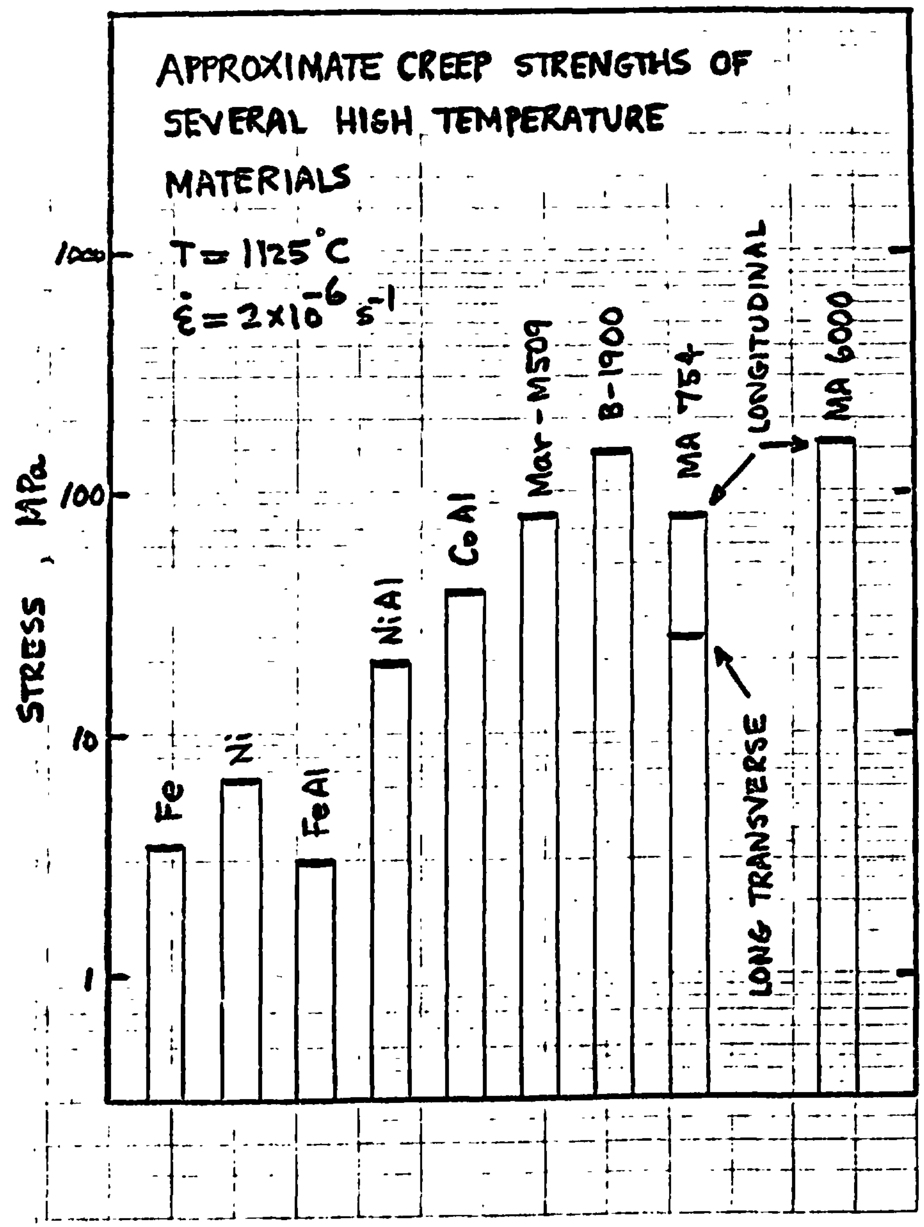
High temperature compression tests have been conducted (at Lewis and Stanford) on polycrystalline NiAl and CoAl made by hot extruding atomized powders. The stress-strain-strain rate characteristics of these materials are described, with particular reference to the possible rate limiting mechanisms for flow. The mechanical data strongly suggest that some kind of lattice friction makes an important contribution to the high temperature strength. The composition dependence of the high temperature strength is also shown. The strength of CoAl depends strongly on deviation from stoichiometry.

The dislocation structures found in both as-extruded and as-deformed samples of CoAl have been studied using TEM. Extensive dislocation networks and very coarse subgrains are found in the as-extruded material. These features are also found in the deformed material, together with additional isolated dislocations. The Burgers vectors of some of the dislocations have been determined to be $a\langle 100\rangle$ and $a\langle 110\rangle$. These dislocations provide sufficient slip systems for general deformation. The scale of the dislocation substructure is much coarser than one would expect for a metal deformed at the same stress. This strongly suggests lattice friction as an important factor in the high temperature strength.



CsCl (B2) structure
(FeAl, NiAl, CoAl)

POSSIBLE SLIP VECTORS:
⟨100⟩, ⟨110⟩ ⟨111⟩



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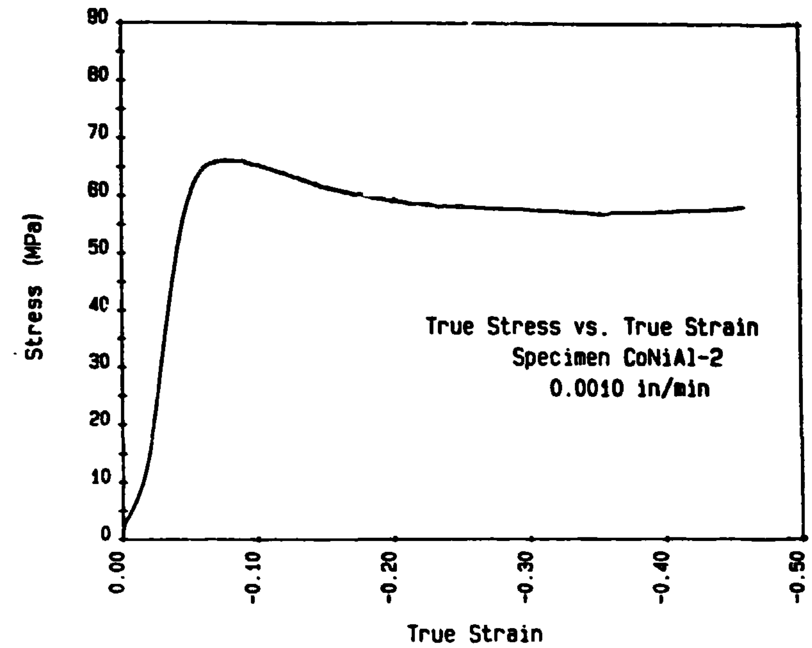

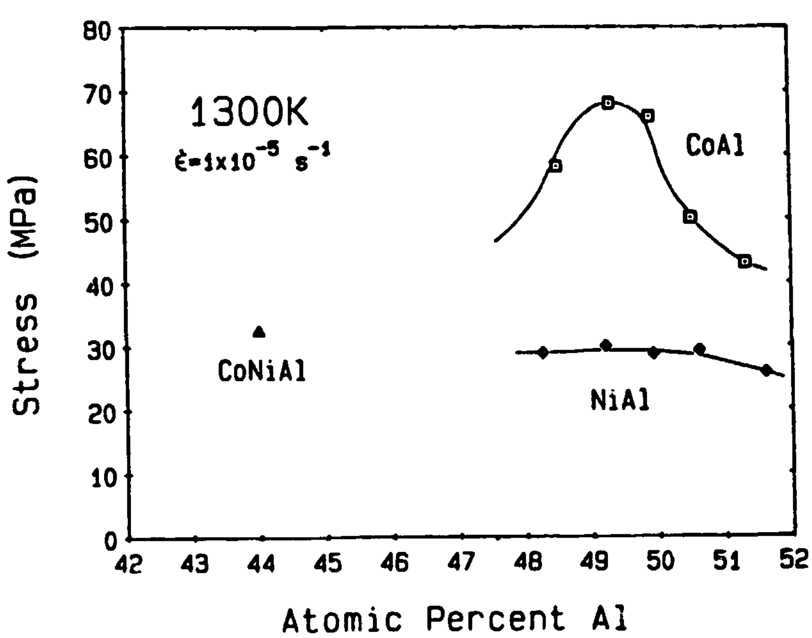

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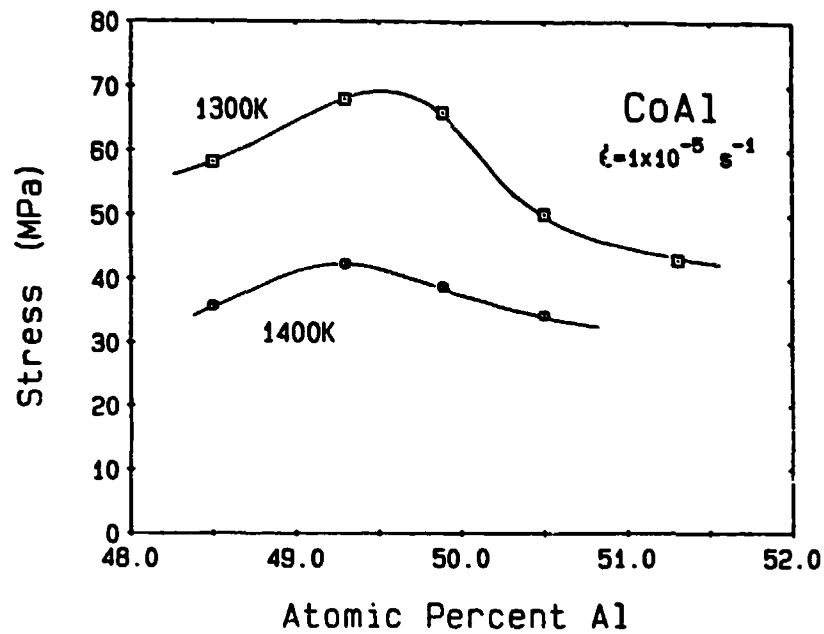

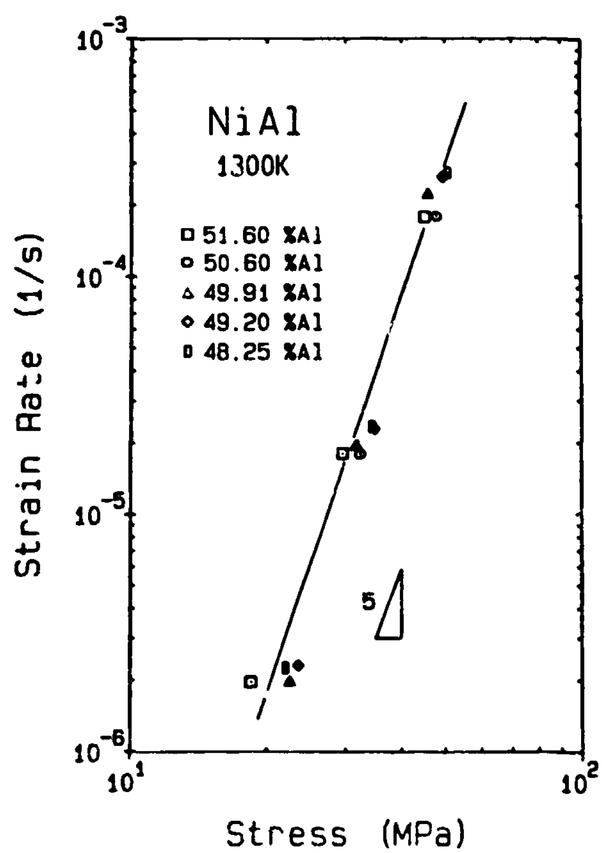

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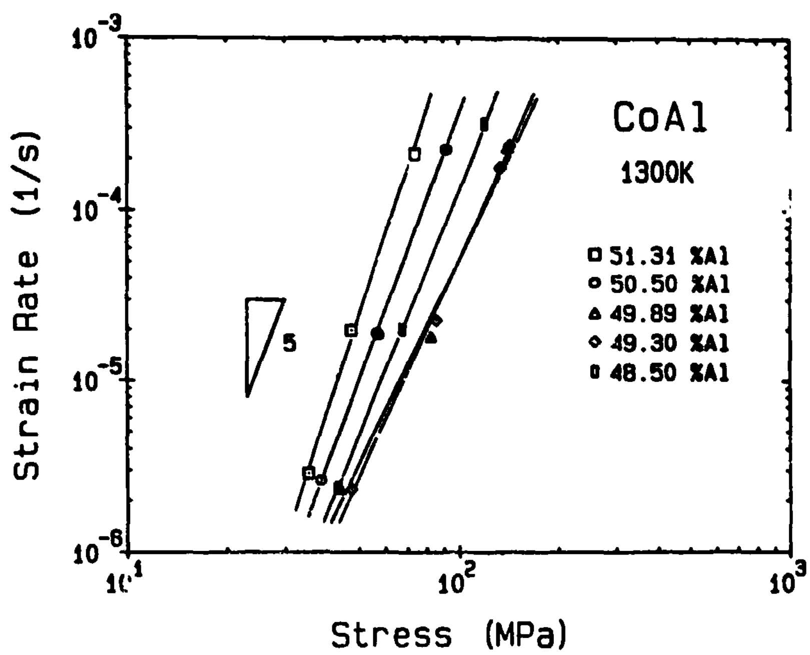
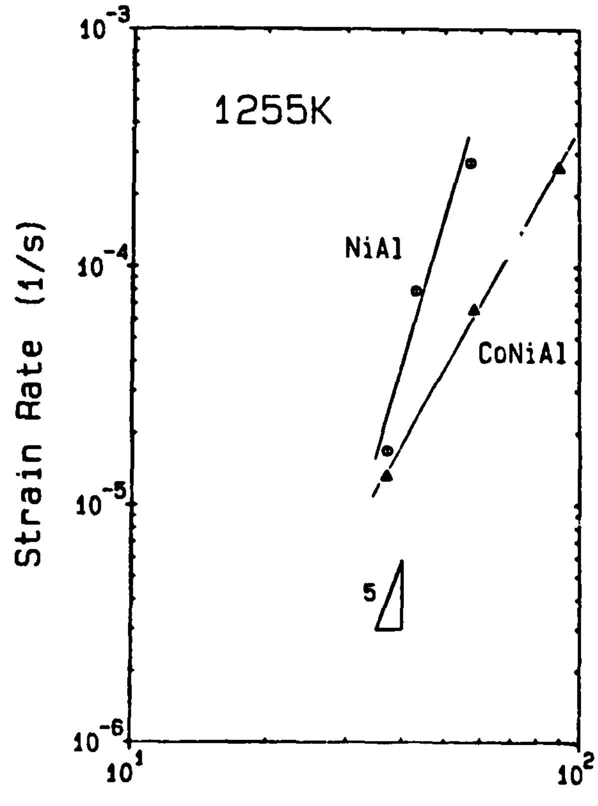
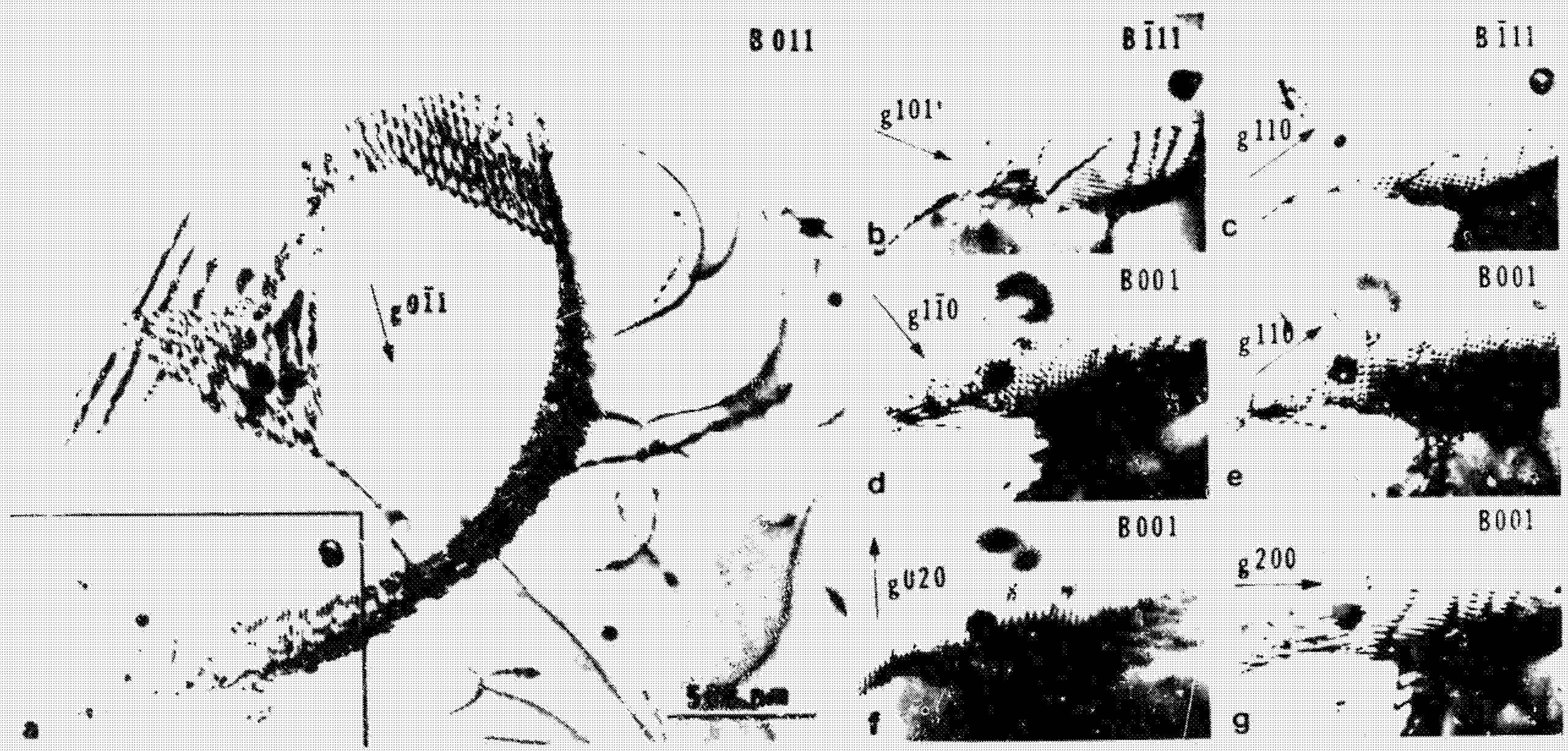
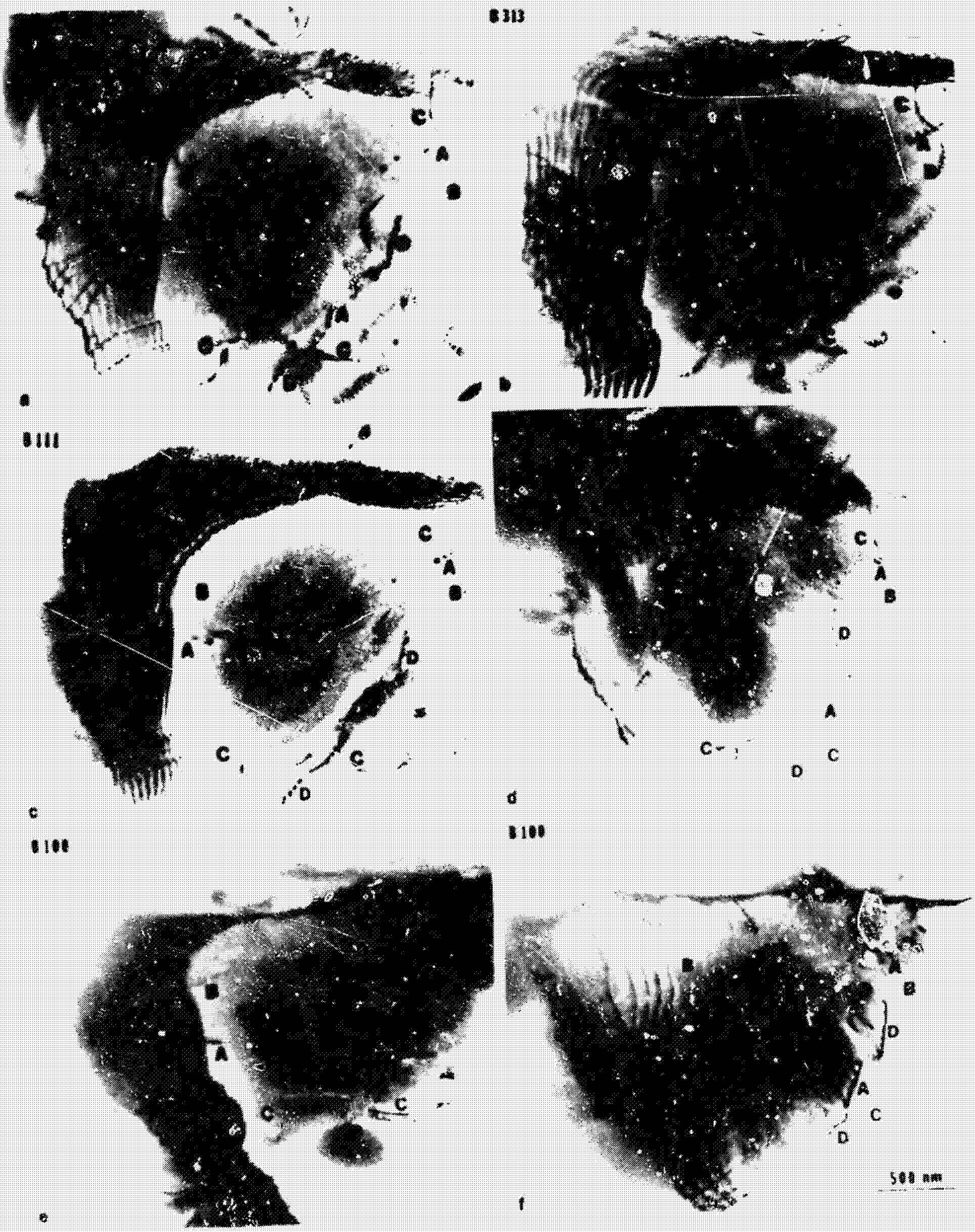


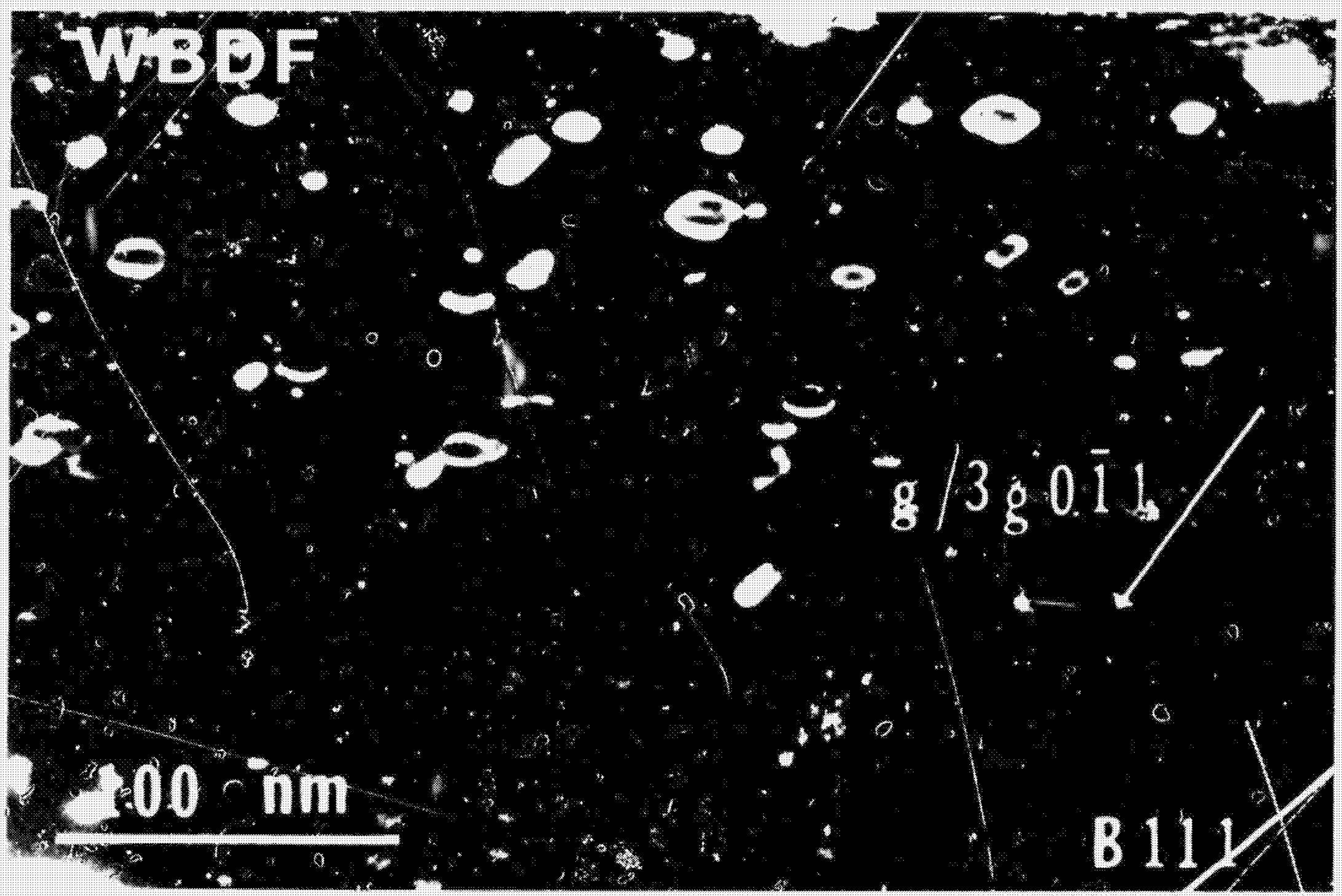




49.3Al-50.7Co
as extruded



49.3Al-50.7Co, as deformed
(1300K, 2×10^{-5} s^{-1}, 9.5%)



48.5Al-51.5Co, as deformed
(1300K, 2×10^{-5} s^{-1}, 9.3%)

N83 11303

THE USE OF THE PUCOT FOR ELASTIC MODULUS MEASUREMENTS ON INTERMATALLICS AT HIGH TEMPERATURES

Alan Wolfenden
Texas A&M University
Mechanical Engineering Department
College Station, Texas 77843

The piezoelectric ultrasonic composite oscillator technique (PUCOT) has proved to be highly successful for research on mechanical properties/microstructure relations in a variety of materials. In the joint research with NASA the technique is being applied to measurements of the elastic constants of the iron aluminides in the temperature ranges 300 to 1700 K (CoAl and NiAl) and 300 to 1500 K (FeAl). The PUCOT consists of piezoelectric quartz drive (D) and gauge (G) crystals to excite longitudinal or torsional ultrasonic (80 kHz) resonant stress waves in the specimen (S) and alumina spacer rod (Q) of appropriate resonant lengths. The resonant system is driven by a closed-loop oscillator which maintains a constant gauge voltage and hence constant strain amplitude in the specimen. While the specimen is heated at 20 K/h the resonant period τ_{DGQS} is measured continuously. The elastic moduli are calculated from these values of τ_{DGQS} and accurate determinations of specimen length. The technique will be described and some results given.

PIEZOELECTRIC
ULTRASONIC
COMPOSITE
OSCILLATOR
TECHNIQUE

PUCOT

Internal Friction Or Mechanical Damping

Ability Of Materials To Dissipate Vibrational Energy

$Q^{-1}(S) = \tan\phi = \phi = \delta/\pi = \Delta W/2\pi W \quad (\phi \ll 1)$

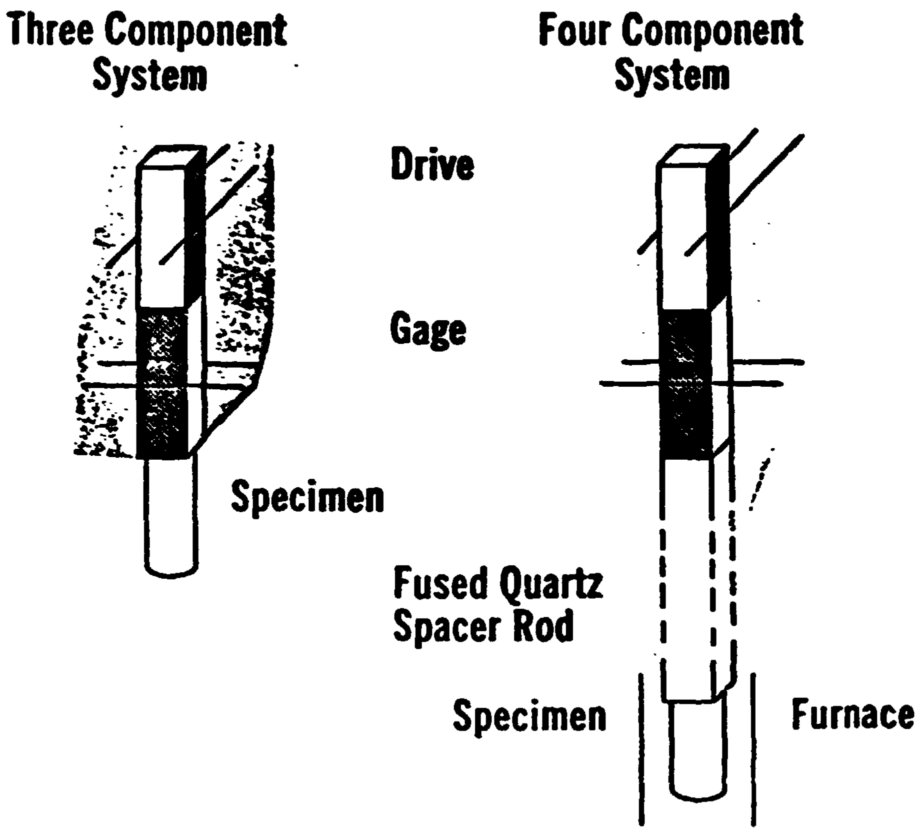
ϕ = Loss Angle (Strain lags Stress)

δ = Logarithmic Decrement

ΔW = Energy Dissipated/Cycle

W = Maximum Stored Energy/Unit Volume

For Measuring:	Mechanical Damping Q^{-1} Internal Friction
	Young's Modulus E
For Monitoring:	Metallurgical Changes



Record: Vd

Vg

τ (DGS) Or τ (DGQS)

T

Use PUCOT Equations To Get: E

Q^{-1}

ϵ

Pucot Equations

$$\ell = (1/2f)\ \sqrt{(E/\rho)} = \lambda/2$$

$$\tau(S) = m(S)^{1/2}\ \tau(DG)\tau(DGS)/A$$

$$A = \{\tau(DG)^2 m(DGS) - \tau(DGS)^2 m(DG)\}^{1/2}$$

$$E = 4\ \rho\ell^2/\tau(S)^2$$

$$Q^{-1}(S) = \{(2/m(S)C_m)\ (N\tau(S)/\pi)^2\}\ Vd/Vg$$

$$\epsilon_{11} = \{(C_m\ \pi\ \sqrt{2})/N\lambda\}\ Vg$$

Pucot Equations

$$\ell = (1/2f)\ \sqrt{(E/\rho)} = \lambda/2$$

$$\tau(S) = m(S)^{1/2}\ \tau(DGQ)\tau(DGQS)/A$$

$$A = \{\tau(DGQ)^2 m(DGQS) - \tau(DGQS)^2 m(DGQ)\}^{1/2}$$

$$E = 4\rho\ell^2/\tau(S)^2$$

$$Q^{-1}(S) = \{(2/m(S)C_m)\ (N\tau(S)/\pi)^2\}\ Vd/Vg$$

$$\epsilon_{11} = \{(C_m\ \pi\ \sqrt{2})/N\lambda\}\ Vg$$

Materials Studied With The PUCOT

Cu_3Au: Order-Disorder Process

Ni-25 a/o Co: Order-Disorder, Magnetic Transformations Near Curie Point

Au-Ag: Order-Disorder

Hg-Sn-Ag: Phase Changes

Pb in Cu-Zn: Melting/Redistribution Of G.B. Frecipitates

Fe: Magnetic Transformations Near Curie Point

Ni: Magnetic Transformations Near Curie Point

Mn-Cu: Precipitation Process

$Fe_{80}B_{20}$: Young's Modulus

NaCl, KCl, LiF, CaF_2: Electro-Mechanical Coupling Of Dislocations

Steels For Turbine Blades: Amplitude Dependence Of Damping

Ti-6V-4 Al: Damping At Low Strain Amplitudes

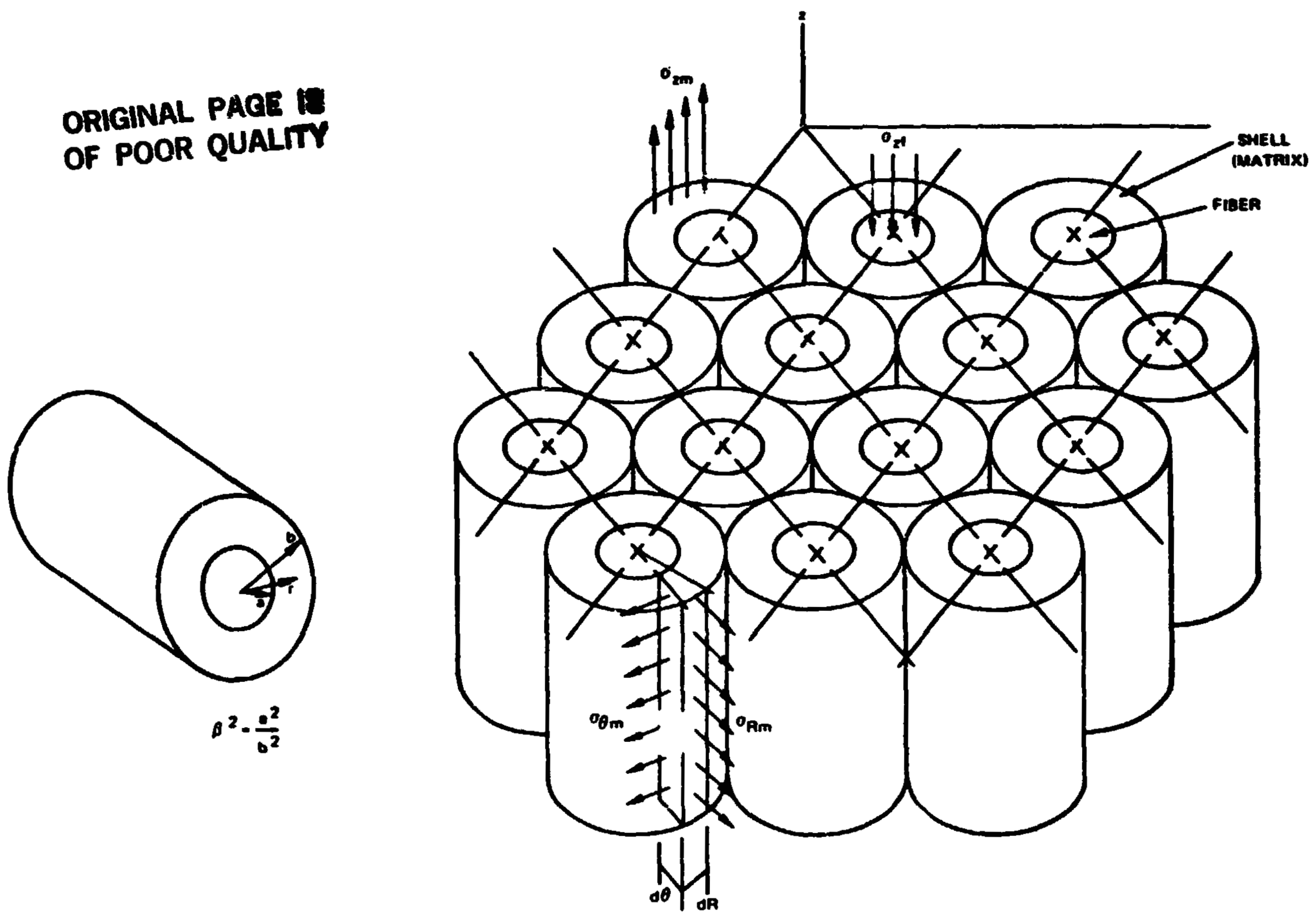
omIT

THERMAL STRAIN MODELING OF IRON-BASE EUTECTICS

David D. Pearson
United Technologies Research Center
East Hartford, Connecticut 06108

Considerable interest has been generated in aligned eutectics as high temperature structural materials. These materials generally are composed of high strength, high modulus fibers or plates in a lower strength lower modulus metal matrix. As a result, the potential exists for the development of large internal stresses due to differences in thermal expansion behavior and modulus of the constituent phases. This condition can cause problems in thermal cycling of these materials. The Fe MnCr-M_7C_3 eutectics are attractive candidates as low cost, high strength materials but little is known about their thermal cyclic behavior. The present talk will consist of an analysis of the thermal strains which could be generated and an outline of experiments performed to characterize thermal effects.

SHELL MODEL FOR FIBER REINFORCED COMPOSITE



EQUATIONS FOR ELASTIC STRESSES DEVELOPED IN INITIALLY STRESS-FREE γFE - M_7C_3 ON COOLING

$$\sigma_{mz} = \frac{V_f E_f E_m\{T[\alpha_f(T)-\alpha_m(T)]-T_H[\alpha_f(T_H)-\alpha_m(T_H)]\} + 2\,V_f P\left[\frac{\nu_m E_f \beta^2}{1-\beta^2} + \nu_f E_m\right]}{V_f E_m + (1-V_f)E_f} \qquad (A9)$$

$$\sigma_{fz} = \frac{-(1-V_f)E_f E_m\{T[\alpha_f(T)-\alpha_m(T)]-T_H[\alpha_f(T_H)-\alpha_m(T_H)]\}-2(1-V_f)P\left[\frac{\nu_m E_f \beta^2}{1-\beta^2} + \nu_f E_m\right]}{V_f E_m + (1-V_f)E_f} \qquad (A10)$$

EUTECTIC EQUILIBRIA IN THE QUATERNARY SYSTEM Fe-Cr-Mn-C

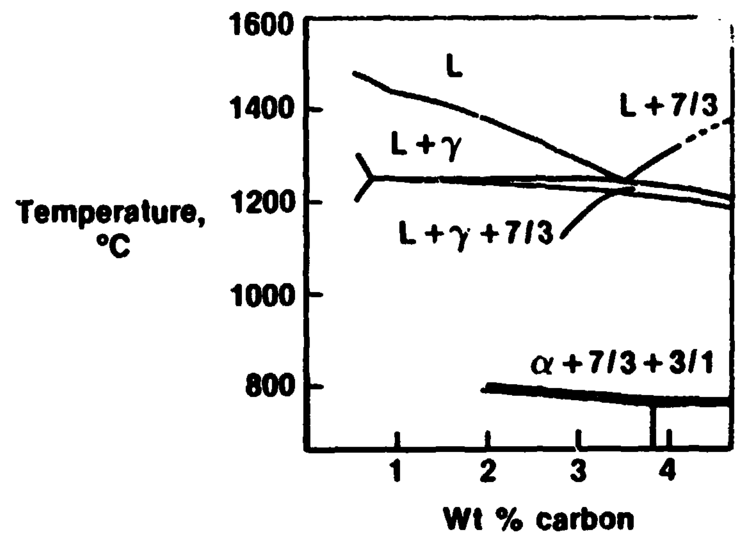
H. Nowotny and S. Wayne
University of Connecticut
Storrs, Connecticut 06268

and

J. C. Schuster
University of Vienna
Vienna, Austria

A challenge exists for wider application of low cost iron-base alloys in the extreme conditions of high temperature, hot corrosion, and high stress experienced in gas turbines. In response to this challenge the constitution of the quaternary system, Fe-Cr-Mn-C and to a lesser extent the quinary system, Fe-Cr-Mn-Al-C were examined for in situ composite alloy candidates. Multivariant eutectic compositions were determined from phase equilibria studies wherein M_7C_3 carbides, present as approximately 30% by volume formed from the melt within gamma iron. An extended field of the hexagonal carbide, $(Cr, Fe, Mn)_7C_3$, was found without undergoing transformation to the orthorhombic structure. Increasing stability for this carbide was found for higher ratios of Cr/Fe + Cr + Mn. Aluminum additions were found to promote a ferritic matrix while manganese favored the desired gamma austenitic matrix. In co-existence with the matrix phase chromium enters preferentially the carbide phase while manganese distributes equally between the gamma matrix and the M_7C_3 carbide. The composition and lattice parameters of the carbide and matrix phases were determined to establish their respective stabilities.

Fe-Cr-C ISOPLETH AT 17% Cr



MATRIX AND CARBIDE COMPOSITION

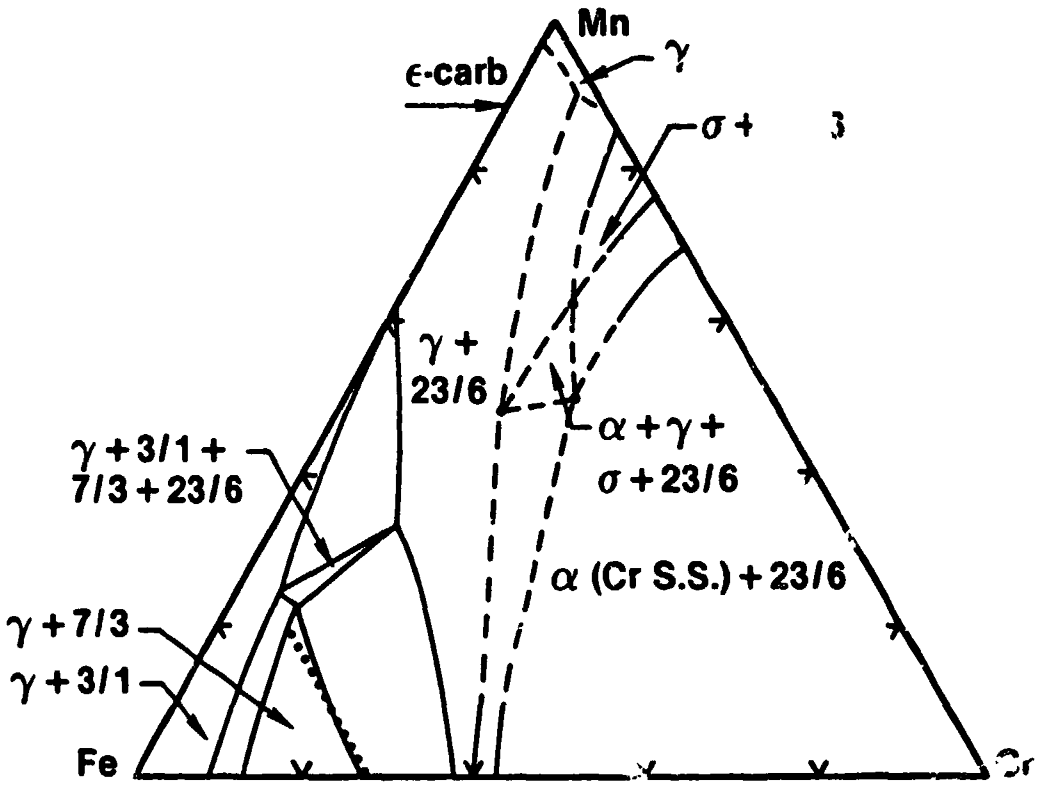
Aligned Fe-Mn-Cr-C alloys

Wt % alloy composition Nominal				Wt % matrix composition			
Fe	Cr	Mn	C	Fe	Cr	Mn	C
66.8	20	10	3.2	80.1	10.6	9.3	0
66.8	15	15	3.2	75.6	7.7	16.7	0
66.7	10	20	3.3	76.3	5.1	18.6	0

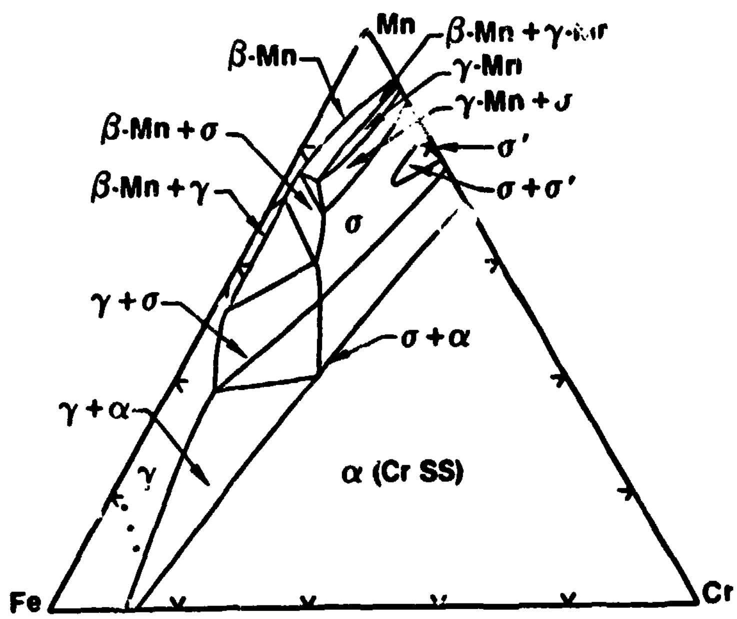
Wt % carbide composition				M_7C_3	ρ (gm/cm^3)
Fe	Cr	Mn	C		
35.2	45.2	10.2	10.7	$Cr_{3.6}Fe_{2.6}Mn_{0.8}C_3$	7.27
30.8	36.6	14.6	11.9	$Cr_{3.3}Fe_{2.5}Mn_{1.2}C_3$	7.3
3b.6	25.7	20.1	13.7	$Cr_{2.2}Fe_{3.1}Mn_{1.7}C_3$	7.4

Fe-Mn-Cr-C AT 1000°C

~ 3 weight % carbon



Fe-Mn-Cr AT 1000°C



(Å) of (23, 6); (3, 1) and (7, 3) carbides

Fe-Mn-C, Fe-Cr-C, and Mn-Cr-C alloys

Composition and constituents	Lattice parameters (Å)					
	(23, 6)	(3, 1)			(7, 3)	
Fe-Mn-C	a	a	b	c	a	c
$Fe_{\sim 11.5}Mn_{\sim 11.5}C_6$	10.535	—	—	—	—	—
$Fe_{\sim 2}Mn_{\sim 1}C$	—	5.04_8	6.74	4.51_3	—	—
$Fe_{2.45}Mn_{4.55}C_3$	—	—	—	—	13.82_0	4.53_2
Fe-Cr-C						
$Fe_{2.2}Cr_{4.8}C_3$	—	—	—	—	14.01	4.48
$Fe_{4.5}Cr_{2.5}C_3$	—	—	—	—	13.92	4.49_2
Mn-Cr-C						
.$n_{\sim 5}Cr_{\sim 2}C_3$	—	—	—	—	13.906	4.536
$Mn_{5.25}Cr_{1.75}C_3$	—	—	—	—	13.88_0	4.535
$Mn_{3.5}Cr_{3.5}C_3$	—	—	—	—	13.90_2	4.556
$Mn_{1.75}Cr_{5.25}C_3$	—	—	—	—	13.97_1	4.534

N83 11305 D23

ELEVATED TEMPERATURE PROPERTIES OF ALIGNED FERROUS EUTECTICS

Franklin D. Lemkey
United Technologies Research Center
East Hartford, Connecticut 06108

Iron base alloys containing aluminum and chromium together with smaller amounts of yttrium and silicon have been of continuing interest for high temperature applications since the 1930's. Solid solution austenitic iron aluminum ternery alloys have a strength advantage over similar ferritic alloys above 1000°F (538°C) but neither type in their present state of development are suitable for service at 1600-1800°F (871-982°C) due to insufficient tensile and creep strength. Strengthening an inherently weak but oxidation resistant solid solution matrix with aligned in situ chromium carbides represents an attractive approach to achieving both surface stability and creep resistance at elevated temperatures.

Aligned microstructures were produced in alloys of approximately 30 wt % (Cr + Mn), about 3 wt % C and the balance Fe consisting of a gamma matrix and the hexagonal carbide $(Cr, Mn, Fe)_7C_3$. The tensile and stress rupture strength to 2000°F (1093°C) of aligned Fe-20 w/t % Cr-10 wt % Mn-3.2 wt % C measured parallel to the carbide reinforcement exceeded those of the strongest iron-nickel superalloys, e.g., CRM-6D developed by Chrysler for automotive turbine application. The cyclic oxidation and sulfidation response of these alloys at elevated temperatures can be markedly improved by aluminum additions.

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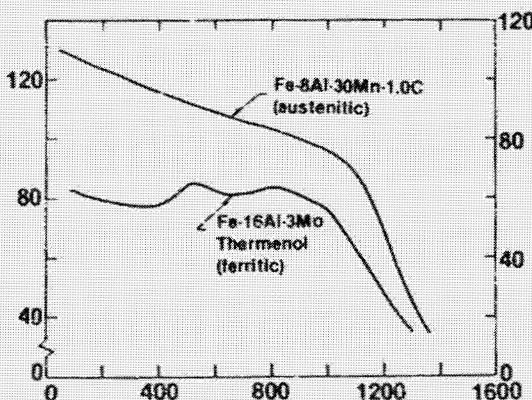
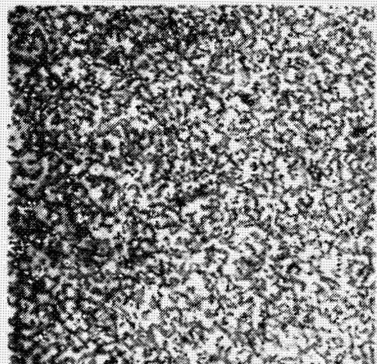
STRESS-TEMPERATURE COMPARISON OF SELECTED AUSTENITIC AND FERRITIC IRON-BASE ALLOYS



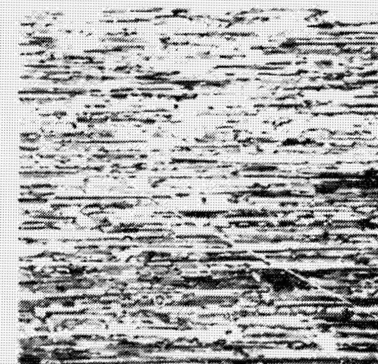
ANISOTROPIC MICROSTRUCTURES

γFe + (Cr, Mn)$_7$ C$_3$

Fe-20Cr-10Mn-3.2C



Transverse 50μm



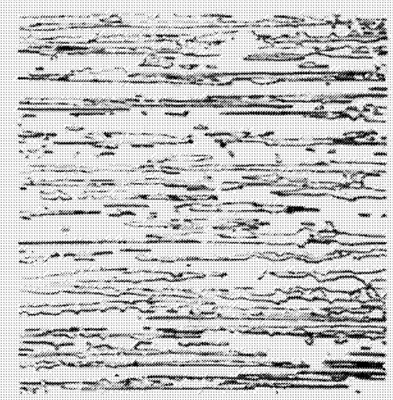
Longitudinal 50μm

ANISOTROPIC MICROSTRUCTURES

$\alpha Fe + Cr_7C_3$

Fe-25Cr-4Al-2.8C

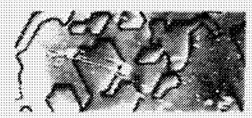
Transverse 50μm



Longitudinal 50μm

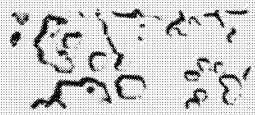
TRANSVERSE MICROSTRUCTURE AND MICROHARDNESS

Cr_7C_3 reinforced iron alloys

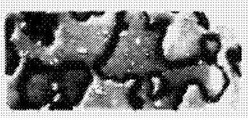


Fe-25Cr-4Al-2.8C
Carbide (VHN) 2003
Matrix (VHN) 209

Fe-20Cr-10Mn-3.2C
17[illegible]4
356



Fe-15Cr-15Mn-3.2C
1550
288

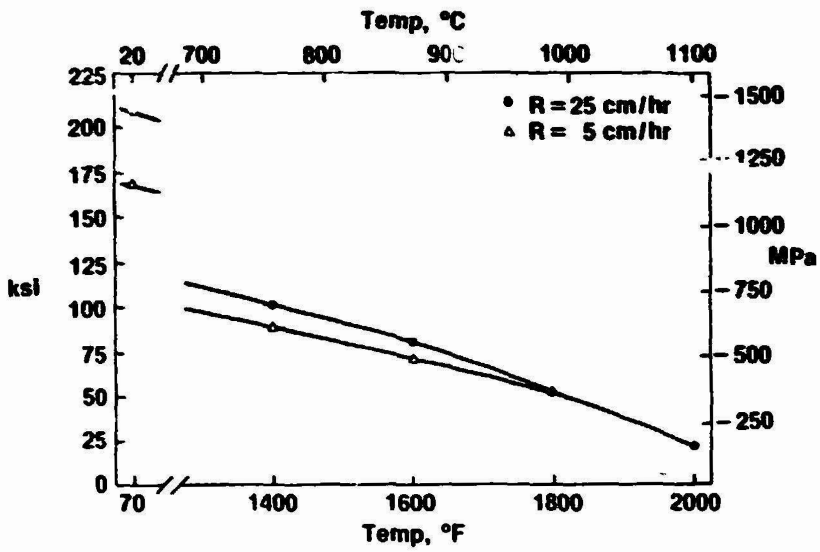


Fe-10Cr-20Mn-3.3C
1467
311

Fe-5Cr-25Mn-3.8C
683
289

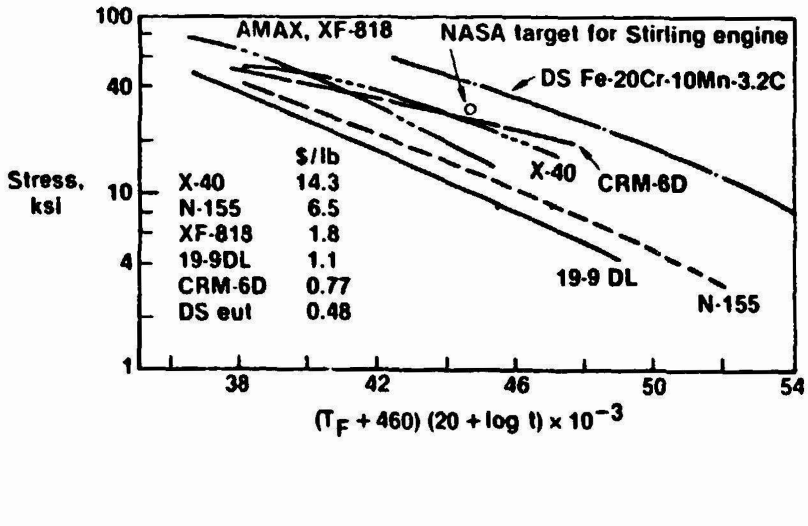
ULTIMATE STRENGTH AS A FUNCTION OF TEMPERATURE

Fe-20 Cr-10 Mn-3.4 C, longitudinal



STRESS RUPTURE STRENGTHS

Fe and Co alloys designed for high temperature usage



After 96 hours cyclic sulfidation testing at 1650°F, 1 mg/cm^2 $NaSO_4$

Composition (w/o)	Δ(mg/cm^2)
Fe-5Cr-25Mn-3.8C	Consumed
Fe-10Cr-20Mn-3.3C	−376.5
Fe-15Cr-15Mn-3.2C	−242.8
Fe-20Cr-10Mn-3.2C	−68.0
Fe-15Cr-15Mn-5Al-2.7C	0.6
Fe-25Cr-4Al-2.8C	0.1

MICROSTRUCTURES OF HOT CORROSION TESTED SPECIMENS

Post 100 hrs, 1650°F, 1 mg/cm^2 Na_2SO_4

Fe-20 w/o Cr-10 w/o Mn-3.2 w/o C

100μm

Fe-25 w/o Cr-4 w/o Al-2.8 w/o C

100μm

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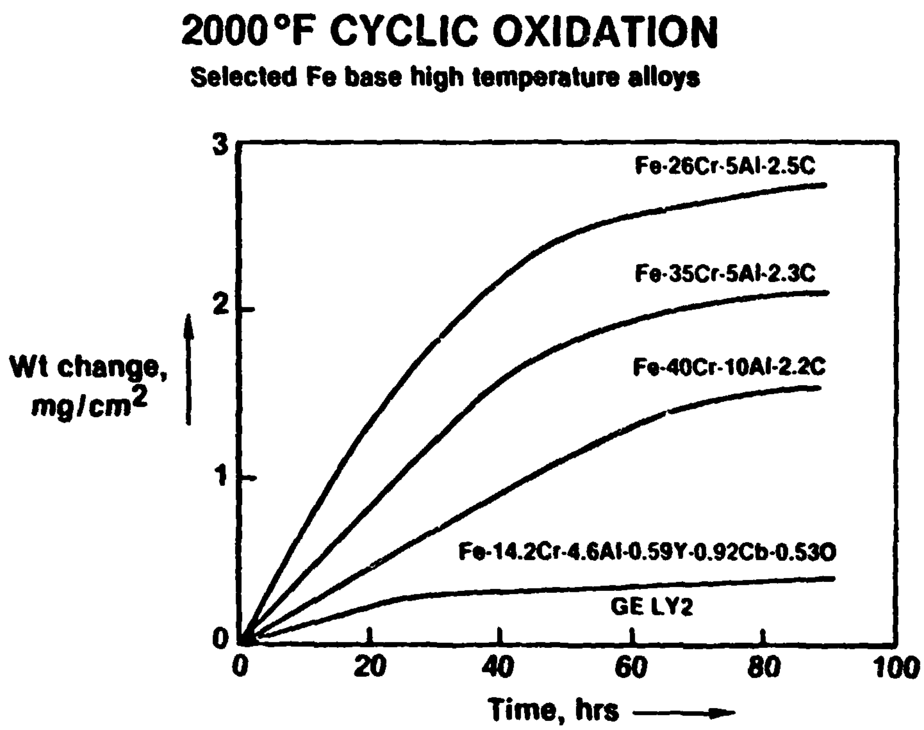


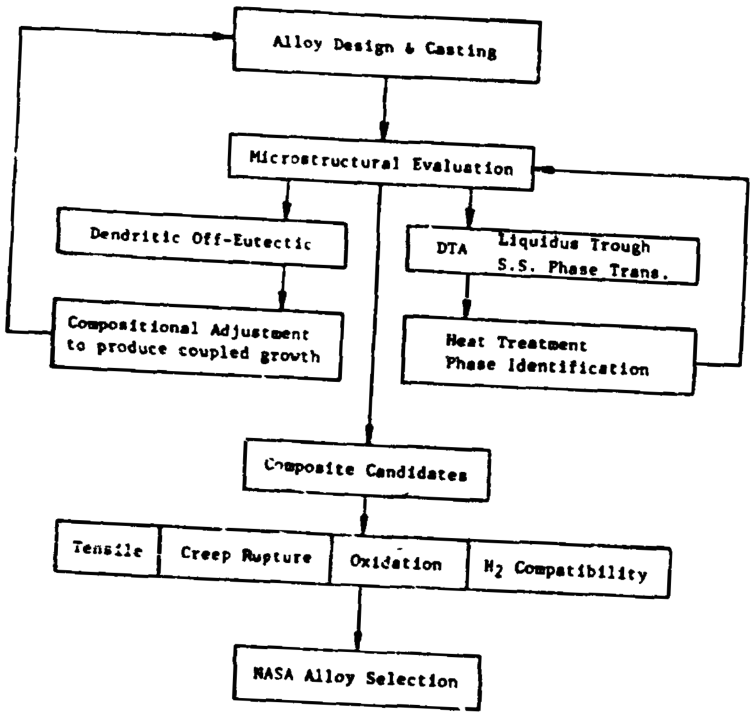
N83 11306 D24

DEVELOPMENT OF CAST FERROUS ALLOYS FOR STIRLING ENGINE APPLICATION

Franklin D. Lemkey
United Technologies Research Center
East Hartford, Connecticut 06108

A need exists for low cost cast ferrous-base alloys that can be used for cylinder and regenerator housing components of the Stirling engine. This alloy must meet the requirements of high strength and thermal fatigue resistance to approximately 1500°F, compatibility and low permeability with hydrogen, good elevated temperature oxidation/corrosion resistance, and contain a minimum of stategic elements. The purpose of this program is to address the development of cast multicomponent ferrous alloys which contain austenitic (γ) matrices reinforced by finely dispersed interdendritic carbides resulting from their combination with other low cost elements such as Mn, Al, C, and N. The phase constituents of over twenty alloy iterations were examined by x-ray diffraction. These alloy candidates were further screened for their tensile and stress rupture strength and surface stability in air at 1450 and 1600°F, respectively. Two alloys, NASAUT 1G (Fe-10Mn-20Cr-1.5C-1.0Si) and NASAUT 4G (Fe-15Mn-12Cr-3Mo-1.5C-1.0Si-1.0Nb), with particular promise towards meeting the program goals,were chosen for more extensive elevated temperature testing. These alloys were found to exhibit nearly equivalent elevated temperature creep strength and oxidation resistance. NASAUT 4G contained additional carbide phases which permit further property evaluation after suitable heat treatment and minor elemental additions of Y, Hf and/or mishmetal. Silicon present in these alloys at the 1 w/o level permitted the achievement of oxide scale adherence to 1600°F without loss of strength (or ductility) as was noted for equivalent additions of aluminum.

Fe-Mn-Cr-Al-C ALLOY ITERATION FLOW CHART



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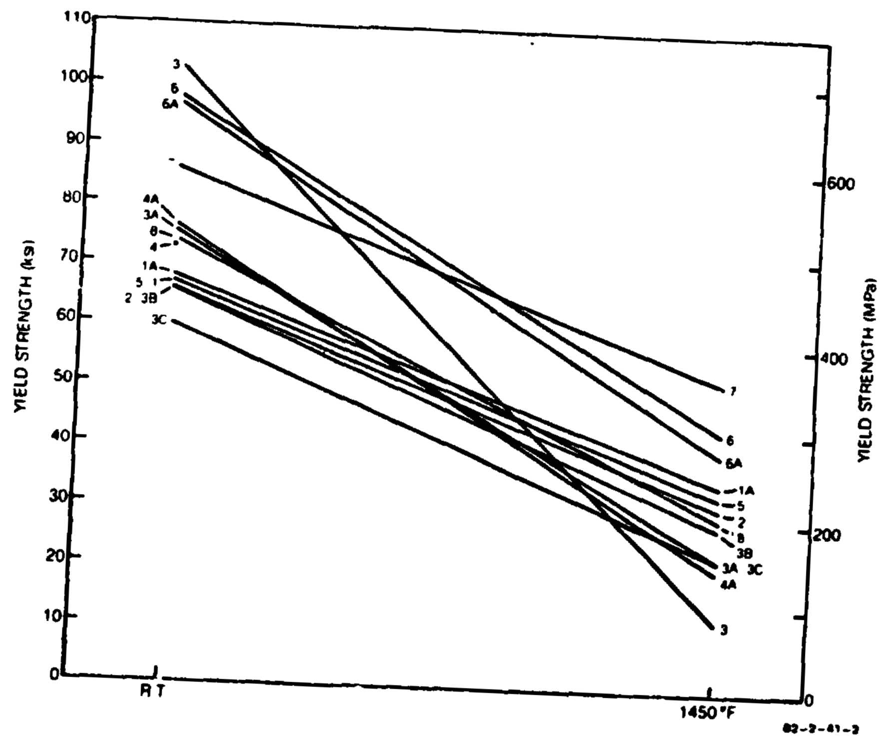
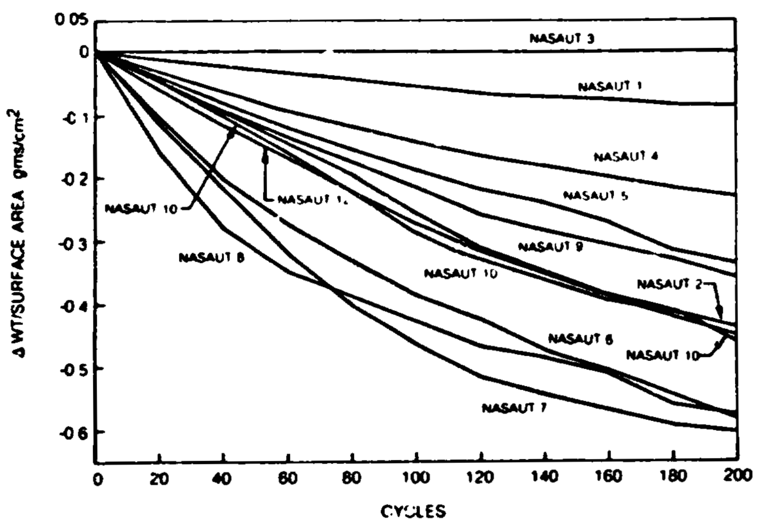
COMPARISON OF ROOM TEMPERATURE AND 1450 °F YIELD STRENGTH OF CANDIDATE ALLOYS WITH 1.5 W/O CARBON

ALLOY NO	Mn	Cr	Mo	Al	C	
	COMPOSITION W/O					
1	10	20	—	—	1.5	—
1A	10	20	—	2.5	1.5	—
2	15	15	—	—	1.5	—
3	10	15	—	5	1.5	—
3A	15	15	—	5	1.5	—
3B	10	15	—	5	1.5	5 Ni
3C	10	15	—	5	1.5	5 Cu
4	15	12	3	—	1.5	—
4A	15	12	3	5	1.5	—
5	15	10	4	2	1.5	—
6	10	—	10	2	1.5	—
6A	10	—	10	5	1.5	—
7	15	5	10	—	1.5	—
8	10	—	4	2	1.5	—
9	15	10	—	2	—	1 N
10	10	15	—	—	1.0	—
12	15	15	—	2	1.0	1 N



WEIGHT LOSS OF CANDIDATE ALLOYS DURING CYCLIC OXIDATION BETWEEN 1600°F AND ROOM TEMPERATURE

ALLOY NO	COMPOSITION W/O Mn	Cr	Mo	Al	C	N
NASAUT 1	10	20	—	—	1	—
NASAUT 2	15	15	—	—	1	—
NASAUT 3	10	15	—	5	1	—
NASAUT 4	15	12	3	—	1	—
NASAUT 5	15	10	4	2	1	—
NASAUT 6	10	—	10	2	1	—
NASAUT 7	15	5	10	—	1	—
NASAUT 8	10	—	4	2	1	—
NASAUT 9	15	10	—	2	—	1
NASAUT 10	10	15	—	—	1	—
NASAUT 12	15	15	—	2	1	1



Composition of Alloy Candidates
(percentage by weight)

Alloy No.	Mn	Cr	Mo	Al	C	Si	Nb
NASAUT 1	10	20	-	-	1.5	-	-
1A	10	20	-	2.5	1.5	-	-
1B	10	20	-	1.0	1.5	-	-
1C	10	20	-	1.5	1.5	-	-
1D	10	20	-	2.0	1.5	-	-
1E	10	20	-	1.5	1.5	1.0	
1F	10	20	-	2.5	1.5	1.0	
1G	10	20	-	-	1.5	1.0	
1H	10	20	-	0.5	1.5	1.5	
NASAUT 4	15	12	3	-	1.5	-	-
4A	15	12	3	5	1.5		
4B	15	12	-	1.5	1.5	-	1.0
4C	15	12	-	2.5	1.5	-	1.0
4D	15	12	-	3.5	1.5	-	1.0
4E	15	12	3	1.5	1.5	-	1.0
4F	15	12	3	0	1.5	1.0	-
4G	15	12	3	0	1.5	1.0	1.0

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Phase Identification of Modifications to NASAUT 1

Alloy No.	Matrix		Carbides	
	Phase	Lattice Parameter, Å	Phase	Lattice Parameter, Å
NASAUT 1	γ(major)	3.62	$M_{23}C_6$ + M_7C_3	a = 10.555 a = 13.82 c = 4.54
NASAUT 1A	γ(major)	3.629	M_7C_3	a = 13.96_6 c = 4.496
NASAUT 1B	γ(major) α(minor)	3.625 2.866	M_7C_3	a = 13.94_1 c = 4.529
NASAUT 1C	γ(major) α(minor)	3.618 2.853	M_7C_3	a = 13.93_5 c = 4.521
NASAUT 1D	α(major) γ(minor)	2.854 3.608	M_7C_3	a = 13.92_2 c = 4.51
NASAUT 1E	γ(major) α(minor)	3.628 2.867	M_7C_3	a ≈ 13.94 c ≈ 4.529
NASAUT 1F	γ(equal) α(equal)	3.625 2.879	M_7C_3	a ≈ 13.94 c ≈ 4.529
NASAUT 1G	γ(major)	3.58	M_7C_3	a = 13.94_2 c = 4.510
NASAUT 1H	γ(major) α(minor)	3.62	M_7C_3	similar to 1G

Phase Identification of Modifications to NASAUT 4

Alloy No.	Matrix		Carbides	
	Phase	Lattice Parameter, Å	Phase	Lattice Parameter, Å
NASAUT 4	γ(major)	3.60	$M_{23}C_6$	
NASAUT 4A	γ(major) α(minor)	3.66_3 2.86_7	M_7C_3	a = 14.06_0 c = 4.521_6
NASAUT 4B	γ(major) α(v. minor)	3.625 2.875	M_7C_3 + NbC	
NASAUT 4C	γ(major) α(minor)	3.633 2.878	M_7C_3 + NbC	
NASAUT 4D	γ(major) α(minor)	3.633 2.862	M_7C_3 + NbC	
NASAUT 4F	γ(major)	3.62	$M_{23}C_6$ + unknown constituent	
NASAUT 4G	γ(major)	3.61	$M_{23}C_6$ + NbC	a = 10.61_4 a = 4.43_3

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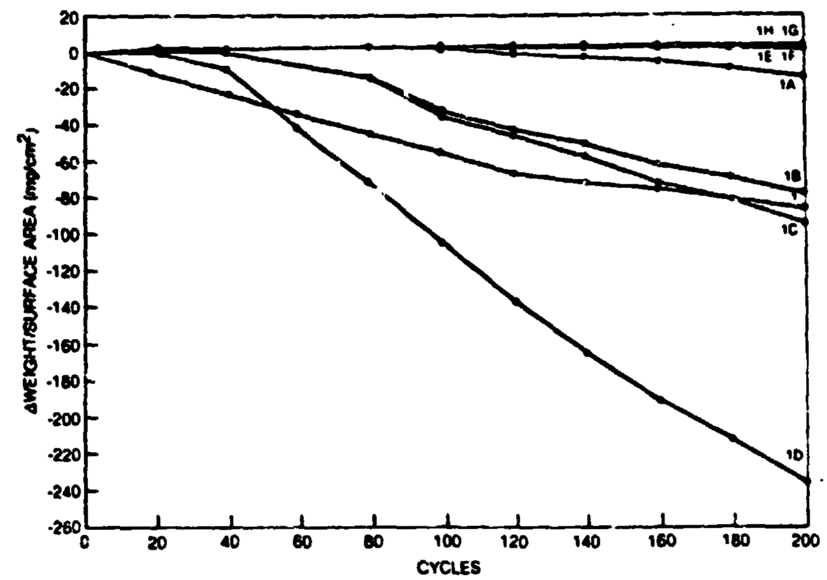
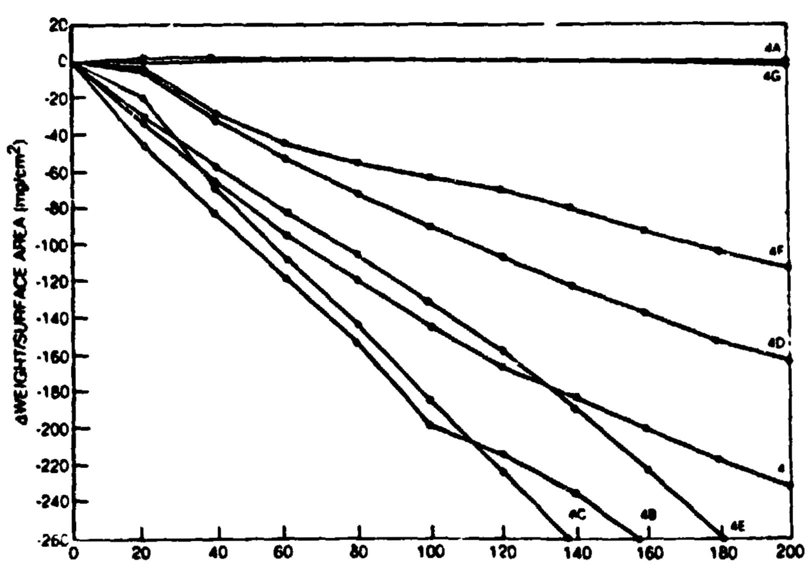
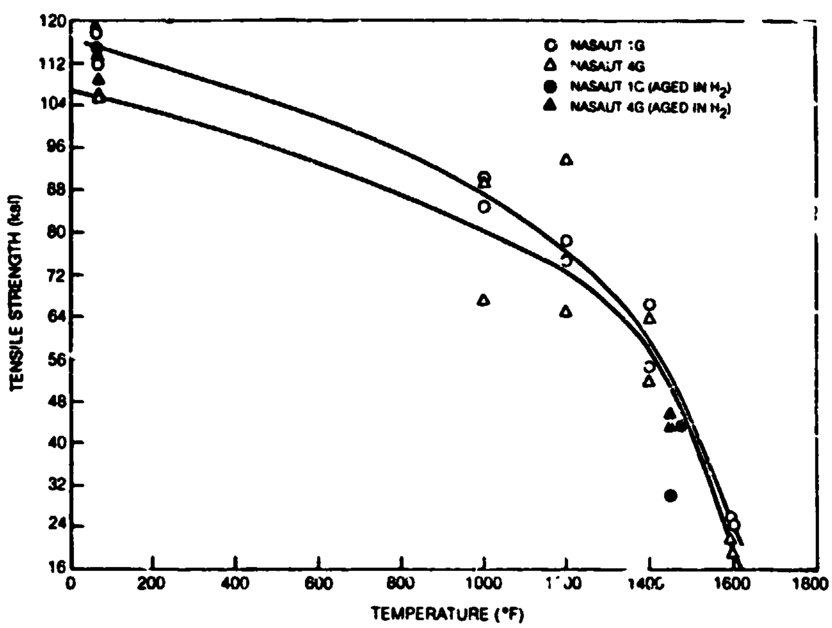


Wt.ght Loss of NASAUT1 Modifications During Cyclic Oxidation Between 1600°F and Room Temperature

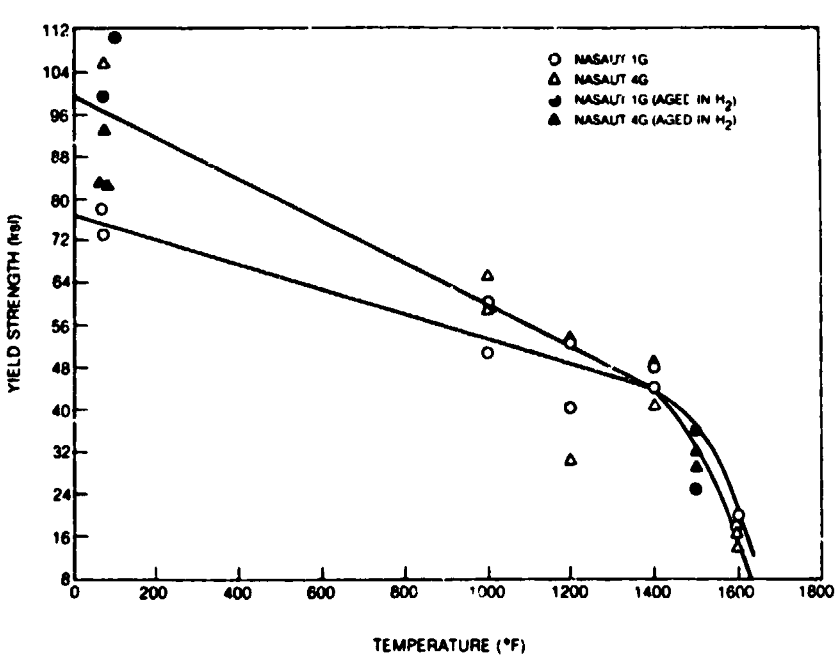


Weight Loss of NASAUT 4 Modifications During Cyclic Oxidation Between 1600°F and Room Temperature

TENSILE STRENGTH VS TEMPERATURE



YIELD STRENGTH VS TEMPERATURE



DEVELOPMENT OF HIGH STRENGTH IRON BASE ALLOYS

Michael J. Woulds
AiResearch Casting Company
Torrance, California

A program was initiated to develop low cost iron base alloys for use in the automotive Stirling engine, as cylinders and regenerator housings.

The goals were: a) the material should be castable, b) stress for a 5000 hour rupture life of 200 MPa (29 Ksi) at 775°C (1427°F)., c) oxidation/corrosion resistance comparable to that of N-155, d) compatibility with hydrogen, e) alloy cost less than or equal to that of 19-9 DL.

The program will be detailed showing how several candidate alloys have been developed that have met or approached the above goals.

The alloys are iron base with minimal reliance on strategic elements. The base composition is Fe-18 Ni-18 Cr-0.5 C-1.0 B-5.0 Mo with additions of Cb or W.

The alloys are austenitic and strengthened by a dispersion of carbides and borides.

Current work will describe efforts in heat treating to increase the ductility of the alloys, as a means of improving the fatigue resistance of the castings.

N83 11307

DEVELOPMENT OF Fe-Mn-Al-X-C ALLOYS

Susan R. Schuon
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

Development of a low cost Cr-free, iron-base alloy for aerospace applications involves both element substitution and enhancement of microstructural strengthening. This work has been divided into three phases: austenite stability, carbide strengthening, and precipitate strengthening. When Mn is substituted for Ni and Al or Si is substituted for Cr, large changes occur in the mechanical and thermal stability of austenite in FeMnAlC alloys. By systematic variation of composition, it has been observed that:

(1) Austenite is not stable at 788° C in Fe25Mn5Al2C or Fe20Mn5Ni5Al2C.

(2) Addition of 10 Mo results in formation of carbides and improved stress rupture life.

(3) High levels of Al lead to low stress rupture probably due to destabilization of austenite.

(4) 788° C tensile properties of cast FeMnAlC alloys are greater or equal to conventional cast stainless steel.

The in situ strength of MC or M_2C (M = Ti, V, Hf, Ta, or Mo) in FeMnAlC alloys has been determined. The high temperature tensile strength depends more on the distribution of carbides than the carbide composition. Precipitation of a high volume percent-ordered phase has been achieved in Fe20Mn10Ni5Al5Ti (1C) alloys. As case, these alloys have a homogeneous austenitic structure. After solutioning at 1100° C for 5 hr followed by aging at 600° C for 16 hr, γ' or a perovskite carbide is precipitated. Overaging occurs at 900° C where η is precipitated. These studies will provide an information base for a low cost, high-strength FeMnAlC alloy for aerospace applications.

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Fe-Mn(Ni)-Al(Si)-X-C

OBJECTIVE: TO DEVELOP A FeMnAlC ALLOY WITH LOW STRATEGIC MATERIAL CONTENT FOR USE IN AEROSPACE APPLICATIONS FROM 650° TO 870° C

ACTION PLAN:

- DETERMINE STABILITY + STRENGTH OF FeMnAlC AS A FUNCTION OF COMPOSITION
- DETERMINE STRENGTH OF MC EUTECTIC CARBIDES IN FeMnAlC AS A FUNCTION OF M
- DETERMINE COMPOSITIONAL LIMITS OF ORDERED PHASE PRECIPITATION

MECHANICAL PROPERTIES OF FeMnAlC ALLOYS

	COMPOSITION								R.T. T.S., MPa		ELONGATION, %	788° C T.S., MPa		ELONGATION, %	172 MPa 788° C SR LIFE, hr
	Fe	Mn	Ni	Mo	Ti	Al	Si	C	UTS	YS	Σ, %	UTS	YS	Σ, %	788° C 172 MPa
1.	BAL	25	-	-	-	5	-	2	924	924	<1	345	290	4	0.1
2.	BAL	20	5	-	-	5	-	2	703	703	<1	214	179	36	0.1
3.	BAL	25	-	10	-	5	-	2	545	545	<1	303	276	13	5.0
4.	BAL	20	5	10	-	5	-	2	876	703	<1	276	248	38	1.35
5.	BAL	20	5	10	1	5	-	2	738	627	<1	303	276	17	4.15
6.	BAL	20	5	10	-	2.5	3	2	448	448	<1	303	225	15	4.3
7.	BAL	25	-	10	-	2.5	3	2	374	374	<1	312	234	12	2.0

EFFECT OF COMPOSITION ON STRENGTH

RT TENSILE STRENGTH

- Ni + LOWER Mn IMPROVE DUCTILITY
- Si DECREASES STRENGTH + DUCTILITY

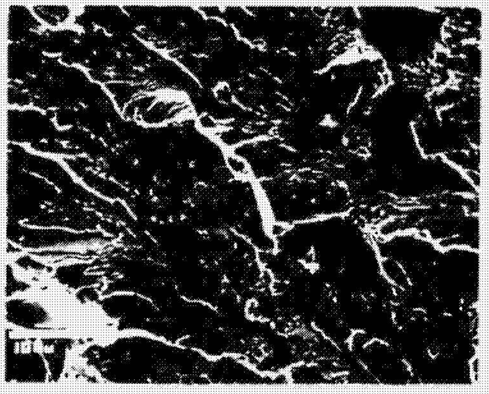
788° C TENSILE STRENGTH

- γ NOT STABLE IN ALLOYS WITHOUT Mo
- Ni, WITHOUT Mo, DECREASES STRENGTH
- STRENGTH OF ALLOYS WITH CARBIDES ≈ 44 ksi

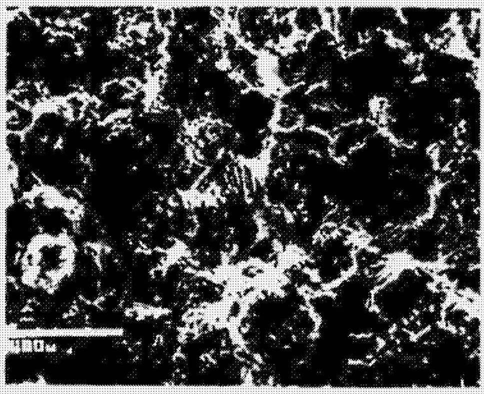
172 MPa, 788° C STRESS RUPTURE LIFE

- CARBIDES RESULT IN IMPROVED S.R. LIFE

EFFECT OF Mo ADDITION ON R.T. FRACTURE OF FeMnAlC ALLOYS



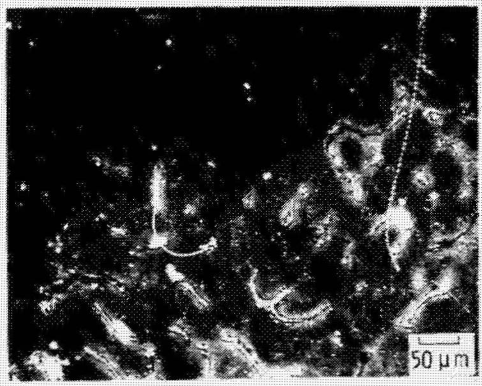
Fe25Mn5Al2C



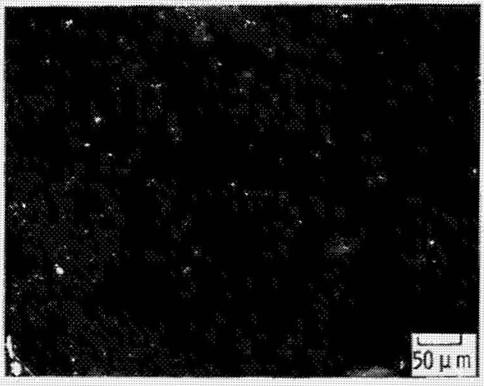
Fe20Mn5Ni5Al(10Mo)2C

CS-82-2267

CARBIDE STRENGTHED FeMnAlC



Fe20Mn5Ni5Al10Mo2C

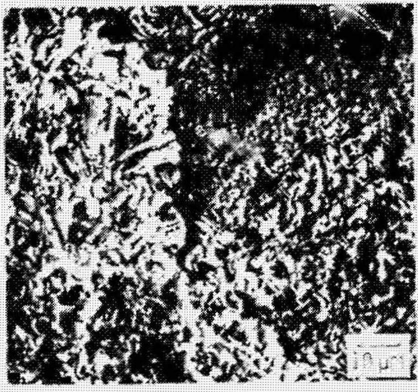


Fe20Mn5Ni2.5Al3Si10Mo2C

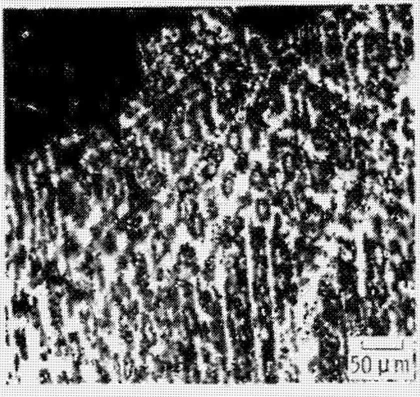
CS-82-2266

EFFECT OF Ni ADDITION ON FRACTURE OF FeMnAlC

STRESS RUPTURE LOAD = 172 MPa;
TEMPERATURE = 788° C



Fe25Mn5Al2C
STRESS INDUCED TRANSFORMATION



Fe20Mn15Ni15Al2C
STRAIN INDUCED TRANSFORMATION

CS-82-2265

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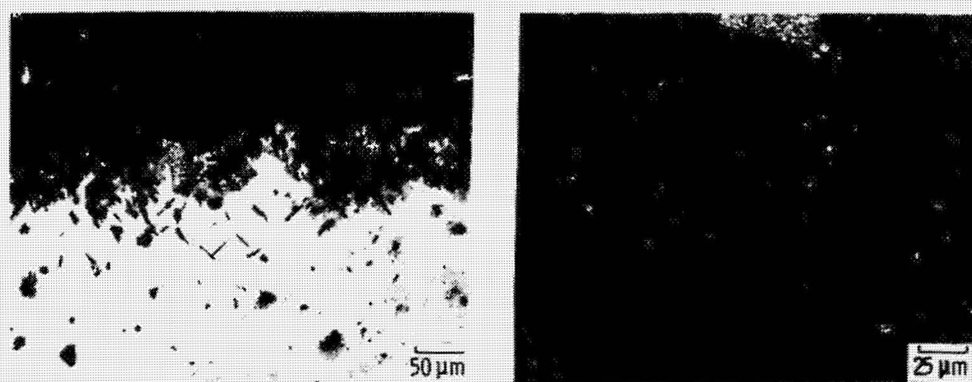
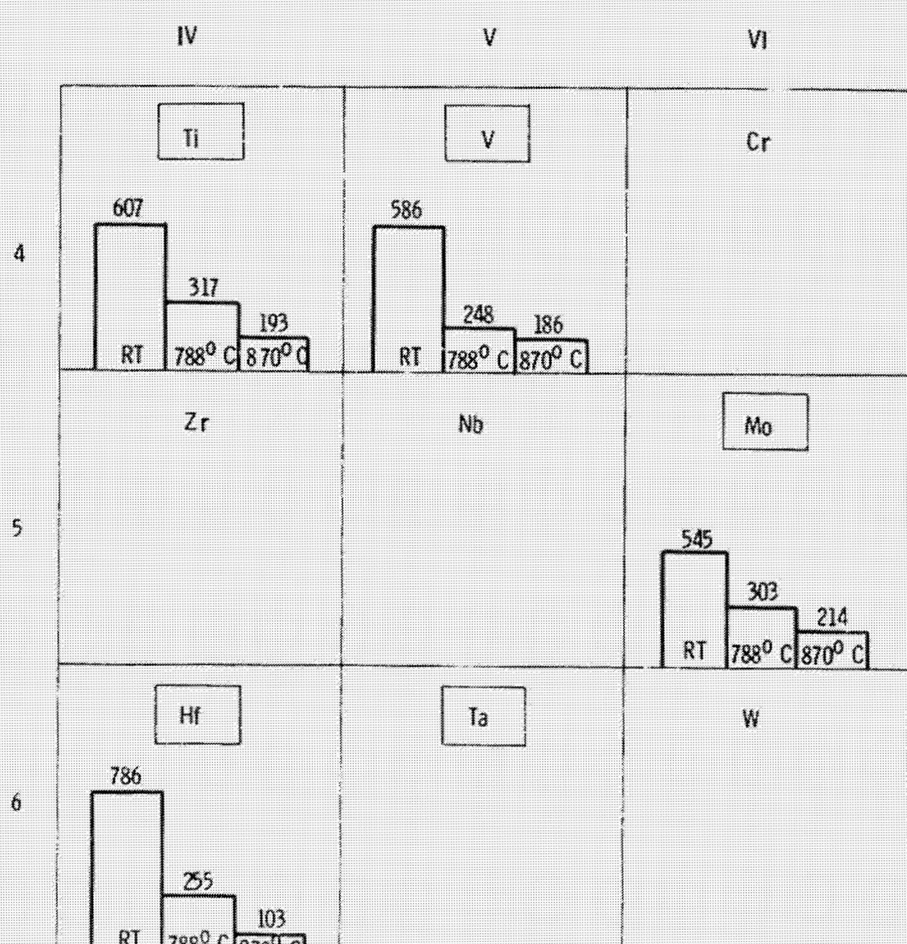
EFFECT OF TEMPERATURE OF THE DEFORMATION INDUCED TRANSFORMATION OF Fe20Mn5Ni5Al2C



R. T. TENSILE

788° C TENSILE

CS-82-2264



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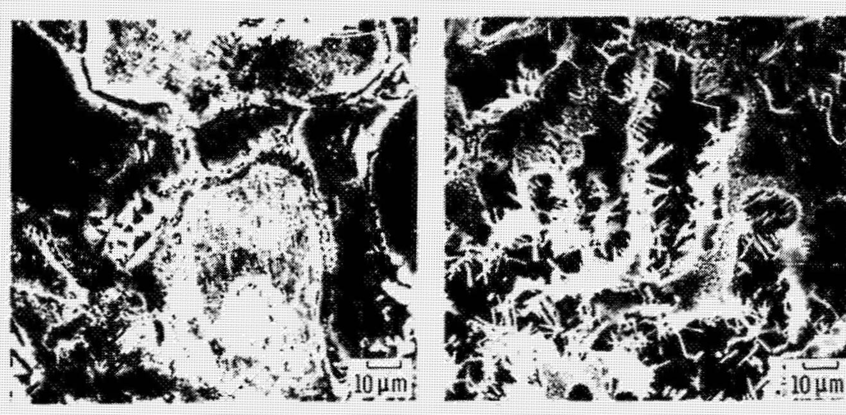
Fe30Ni5Al5Ti

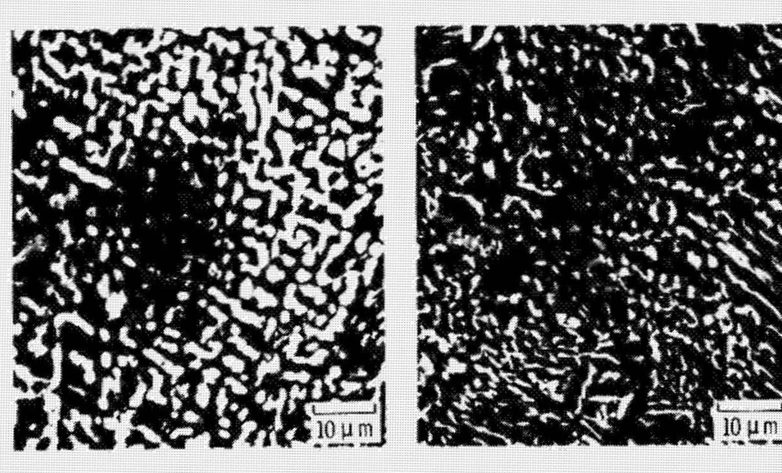


5 hr, 1100° C — 16 hr, 600° C
ORDERED PHASE

5 hr, 1100° C — 16 hr, 900° C
OVERAGED η

CS-82-2268

Fe20Mn10Ni5Al5Ti



5 hr, 1100° C — 16 hr, 600° C
ORDERED PHASE

5 hr, 1100° C — 16 hr, 900° C
OVERAGED η

CS-82-2269

SUMMARY

- γ IS NOT STABLE IN ALLOYS WITHOUT CARBIDES
- Ni IMPROVES DUCTILITY BUT REDUCES STRENGTH
- Si DECREASES R.T. STRENGTH AND DUCTILITY
- Al SEVERELY DECREASES S.R. LIFE
- CARBIDES RESULT IN IMPROVED S.R. LIFE
- STRENGTH IS LESS DEPENDENT ON M IN MC EUTECTIC CARBIDES THAN IN CARBIDE CONTINUITY
- AN ORDERED PHASE (γ' OR PEROVSKITE CARBIDE) CAN BE PRECIPITATED IN CONVENTIONALLY MELTED FeMnNiAl ALLOYS WITH Ni AS LOW AS 10%

N83 11308

SiC or B_4C-B/LOW STRATEGIC ELEMENT CONTENT COMPOSITE

Donald W. Petrasek
National Aeronautics and Space Administration
Lewis Research Center
Cleveland, Ohio

Advancements in materials technologies are needed to provide the aerospace industry with alternate material options in the event of future strategic metal shortages and to optimize performance of engines. Composite materials are promising candidates for such an application. A program is being conducted to determine the potential of SiC and B_4C-B filament reinforced low strategic element iron-base alloy content composites for use at 760° to 870° C (1400° to 1600° F). A limiting factor towards developing this material for high temperature use has been the reaction between the filament and matrix material during fabrication and service which degrades filament strength. A low temperature fabrication process to limit filament/matrix reaction is being developed which involves the use of hollow cathode sputtering to coat the filaments with iron-base alloys of various compositions. An investigation is being conducted to determine the interfacial reaction effects of SiC and B_4C-B filaments with iron-base alloys to develop an understanding of filament/matrix alloy compatibility for 760° to 870° C (1400° to 1600° F) service.

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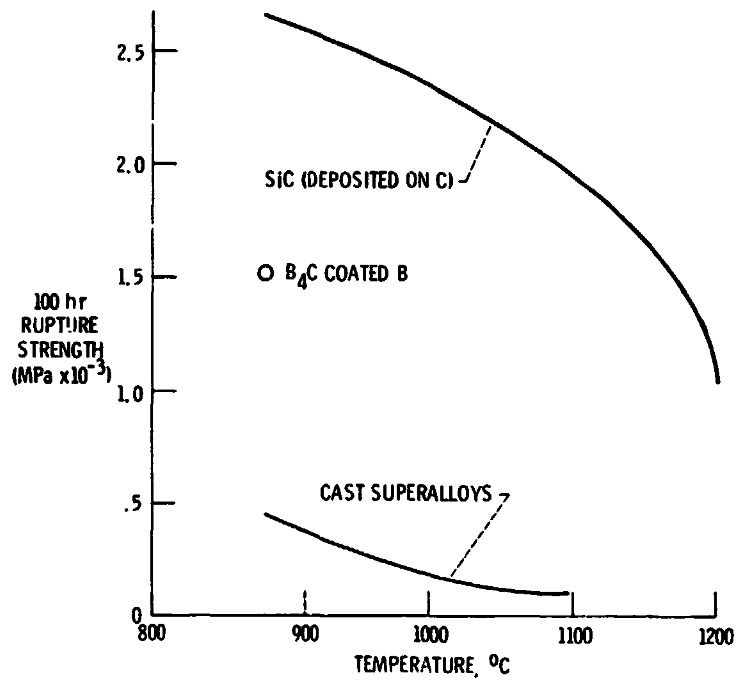
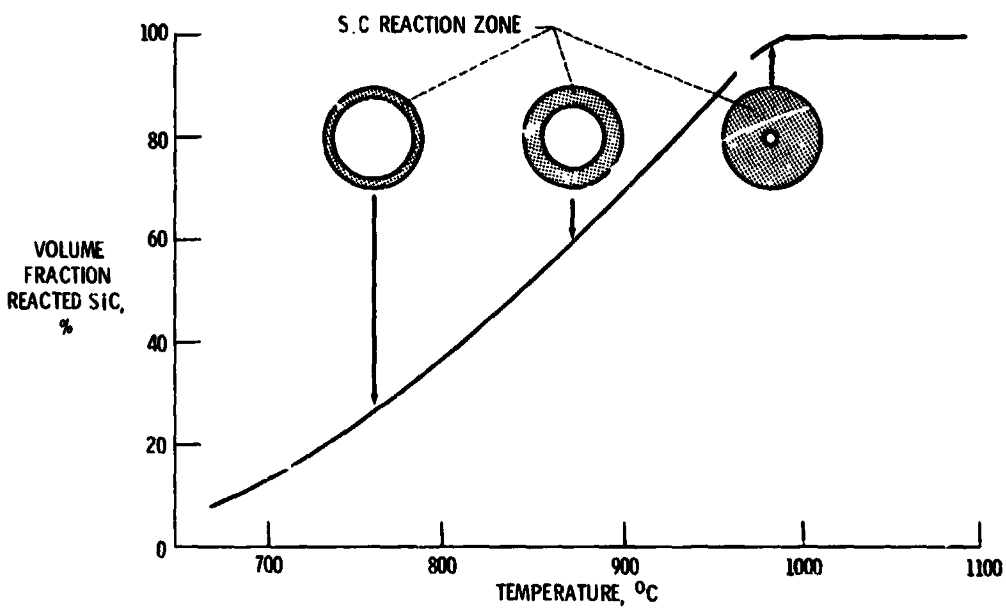
100 hr RUPTURE STRENGTH FOR SiC AND B_4C-B FILAMENTS AND CAST SUPERALLOYS



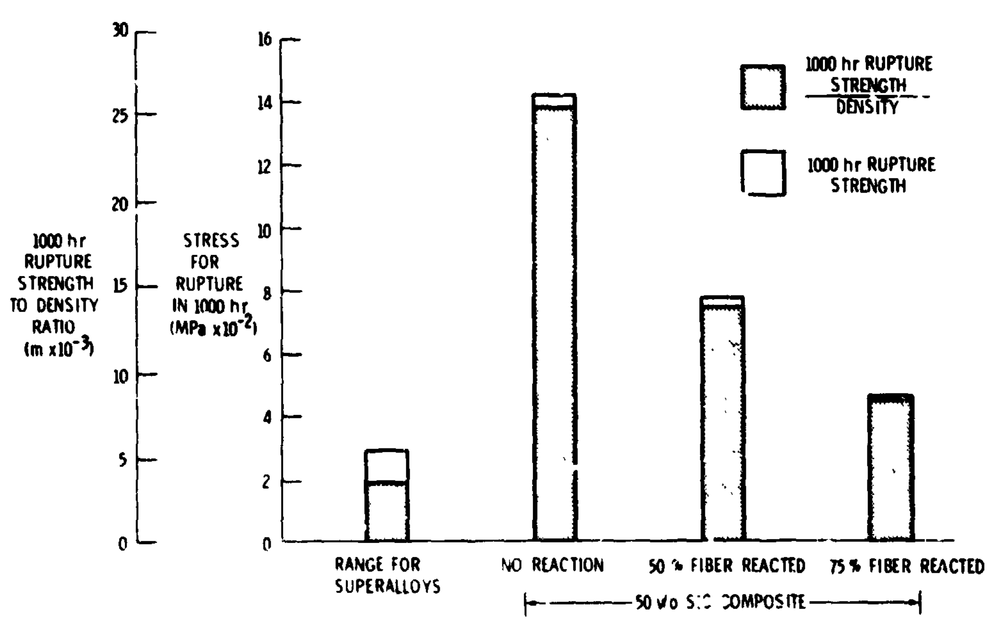
SiC OR B_4C - B/LOW STRATEGIC ELEMENT CONTENT COMPOSITE

- FILAMENT STRENGTH LOSS DUE TO REACTION WITH THE MATRIX IS A POTENTIAL PROBLEM
- LITERATURE DATA INDICATE FILAMENT REACTION AT 980° C OR HIGHER IS SEVERE, AT 760 - 870° C MIXED RESULTS
- CONTROL OF REACTION IS MAJOR FOCUS OF PLANNED PROGRAM, HOWEVER EVEN WITH SOME REACTION A SIGNIFICANT POTENTIAL ADVANTAGE COULD BE OBTAINED FOR THIS COMPOSITE

REACTION OF SiC WITH RENÉ 80 AFTER EXPOSURE FOR 100 hr



POTENTIAL 1000 hr RUPTURE STRENGTH AND 1000 hr RUPTURE STRENGTH TO DENSITY RATIO AT 870° C FOR SiC COMPOSITE COMPARED TO SUPERALLOYS



SiC OR B_4C - B/LOW STRATEGIC ELEMENT CONTENT COMPOSITE

OBJECTIVE:

DETERMINE POTENTIAL OF SiC AND B_4C - B REINFORCED LOW STRATEGIC ELEMENT CONTENT COMPOSITES FOR USE AT 760 - 870° C SERVICE TEMPERATURES FOR TURBINE ENGINE COMPONENTS

JUSTIFICATION:

- REDUCE STRATEGIC ELEMENT CONTENT FOR HOT TURBINE ENGINE COMPONENTS BY SUBSTITUTION OF SiC, B_4C - B AND Fe FOR SCARCE MATERIALS
- REDUCE WEIGHT OF COMPONENTS BY USE OF LOW DENSITY MATERIALS
- INCREASE STRENGTH OF COMPONENT MATERIAL RESULTING IN POTENTIAL LONGER SERVICE LIFE AND REDUCED MAINTENANCE COSTS

SiC OR B_4C - B/LOW STRATEGIC ELEMENT CONTENT COMPOSITE

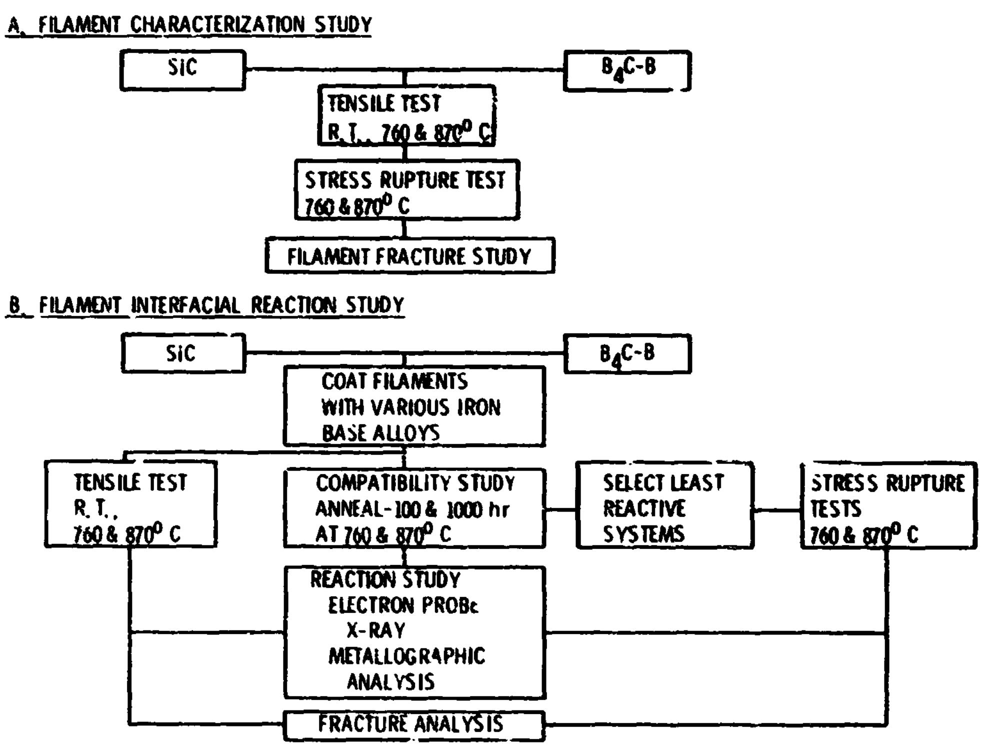
PROBLEM:

- REACTION BETWEEN FILAMENT AND MATRIX DURING FABRICATION AND SERVICE DEGRADES FIBER STRENGTH

APPROACH:

- DEVELOP LOW TEMPERATURE FABRICATION PROCESS TO LIMIT FILAMENT/MATRIX REACTION
- INVESTIGATE THE INTERFACIAL REACTION EFFECTS OF SiC AND B_4C-B FILAMENTS WITH Fe BASE ALLOYS TO DEVELOP AN UNDERSTANDING OF MATRIX ALLOY/FILAMENT COMPATIBILITY FOR 760 - 870° C SERVICE

SiC OR B_4C-B/LOW STRATEGIC ELEMENT CONTENT COMPOSITE



SiC OR B_4C - B/LOW STRATEGIC ELEMENT CONTENT COMPOSITE

STATUS

A. FILAMENT CHARACTERIZATION STUDY

1. TENSILE STRENGTH

SiC	B_4C - B
R.T. - 4385 MPa	R.T. - 4371 MPa
870^0 C - 2985 MPa	870^0 C - 1930 MPa

2. STRESS RUPTURE STRENGTH

SiC	B_4C - B
870^0 C - 100 hr - 2620 MPa	870^0 C - 100 hr - 1551 MPa
- 1000 hr - 2551 MPa	- 1000 hr - 1413 MPa

B. FILAMENT INTERFACIAL REACTION STUDY

CONTRACT AWARDED TO BATTELLE TO DEVELOP A SPUTTERING PROCESS TO COAT SiC AND B_4C - B FILAMENTS WITH VARIOUS IRON BASE ALLOYS. COATED FILAMENTS WILL BE SUPPLIED TO THE GOVERNMENT FOR THIS PHASE OF THE PROGRAM